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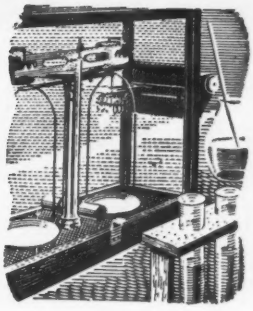

Corrosion Resistance of Some Austenitic Cr–Ni Steels of 18/8/Ti Composition

THE EFFECT OF VARIATION IN CHEMICAL COMPOSITION AND THERMAL TREATMENTS

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SYNOPSIS

The titanium-bearing austenitic chromium–nickel steels of 18/8/Ti type possess excellent intrinsic resistance to corrosion by media of the ' nitric acid ' type, but this resistance may be reduced as the result of heating the material, as in welding, to temperatures in two ranges, viz. (*a*) about 600–750° C and/or (*b*) about 1300° C. The degree of impairment varies with different steels of the same general composition.

The influence of the titanium and silicon content on the susceptibility to impairment of corrosion resistance of the 18/8/Ti steels containing 0·10–0·13% of carbon has been investigated and it is demonstrated that the titanium content is a most important factor.

The influence of the carbon content has also been examined and it is shown that the 18/8/Ti steels having a carbon content not exceeding 0·06% and an adequate titanium/carbon ratio are almost completely immune from susceptibility to both ' low ' and ' high ' temperature impairment of corrosion resistance.

It is concluded that for welded chemical plant items to be exposed to onerous corrosive conditions of the ' nitric acid ' type, the 18/8/Ti steels having a maximum carbon content of 0·06%, with a suitable titanium/carbon ratio, are much superior to those containing 0·10% and above of carbon. 1136

Introduction

ONE OF THE MOST IMPORTANT applications of the austenitic chromium-nickel steels of the 18/8 type containing titanium is for chemical plant fabricated by welding and used for handling nitric acid liquors over a wide range of concentrations and temperatures. Such equipment ranges from storage tanks and pipelines, which may be exposed to the cold acid, to pressure vessels such as autoclaves and ancillary equipment, which may be in contact with nitric acid at elevated temperatures and pressures. The welded joints in such equipment are required to be of a suitable strength and soundness to carry the mechanical loads involved and of adequate corrosion resistance to withstand the chemical conditions and it is fortunate that when suitable precautions are taken, little difficulty is experienced in obtaining welds of a satisfactory standard of quality. The application of a fusion-welding process to these steels may have a detrimental effect on the corrosion resistance of the parent plate material adjacent to the welded joint at those regions which have been heated in the course of the welding operation, a circumstance which is

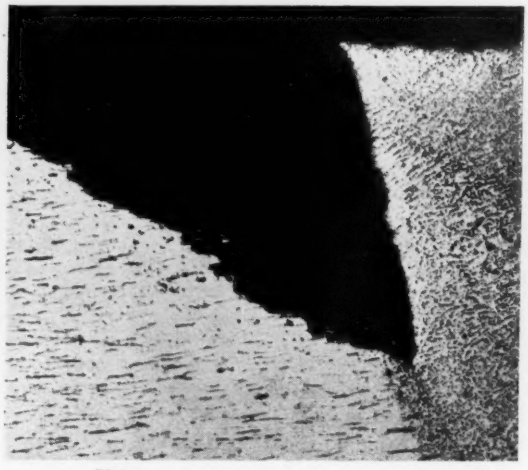

Plate Weld

Fig. 1—Photomicrograph of a section from a welded 18/8 Ti plate after exposure to 60% HNO_3 at 100° C. Note the ' notch ' character of the fissure formed by corrosion wastage × 75

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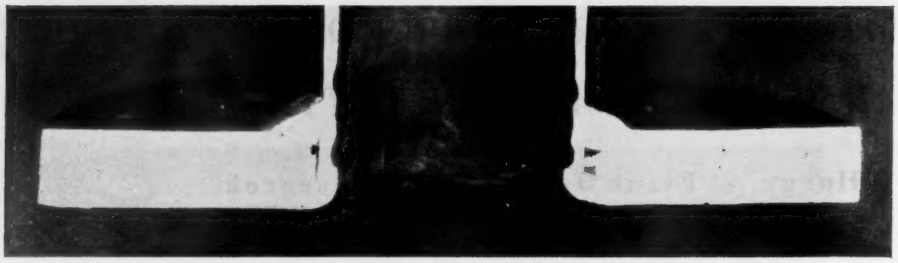

Fig. 2—Longitudinal section of an 18/8 Ti flanged pipe, which failed in service handling warm strong HNO_3. Note the local thinning of the tube material adjacent to (i) the fillet weld at the bore and (ii) the back fillet weld attaching the flange to the pipe Actual size

often of considerable concern to the users of welded chemical plant.

Experience with welded equipment in the 18/8/Ti steels exposed to onerous corrosive conditions of the 'nitric acid' type has shown that impairment of corrosion resistance due to the temperatures attained during the welding operations may occur in two distinct zones, one where the material has been heated to a temperature between about 1300° C. and the melting point of the alloy, and the other where the temperature has been within the range 600–750° C. The former gives rise to the 'fissure' effect shown in Fig. 1 and the latter to the 'grooving' illustrated in Figs. 2 and 3.

Although the corrosion resistance of the 18/8/Ti material which has not been heated in the course of the welding operation may be sufficient to ensure an adequate service life, the reduced resistance to corrosion of the material in the two zones adjacent to welds may determine the useful operating life of welded chemical plant, or may even jeopardize the safe working thereof. For example, the sharp-edged discontinuity which is produced as a result of wastage of the 'fissure' type (*see* Fig. 1) may be a serious stress raiser in the case of a pressure vessel.

The susceptibility to impairment of corrosion resistance at welds of 18/8/Ti steels supplied against a typical ordering specification* has been shown by numerous observations in the field and in the laboratory to vary considerably both as regards the '1300° C' and the '600–750° C' effects. For example, two adjacent plates in a welded vessel on plant service have shown different susceptibilities as regards the 1300° C effect, one exhibiting considerable fissure wastage, whilst the other was apparently unaffected.

The variations in susceptibility of the 18/8/Ti steels to the effect of heating to 600–750° C have been studied previously by the authors and in some work (unpublished) it has been shown that

(i) The 18/8/Ti steels containing 0·10–0·13% C which were examined exhibited very considerable differences in the *degree* of impairment of corrosion resistance (to 'nitric acid' conditions) caused by heating within the range 600–750° C.

(ii) The temperature of heat-treatment corresponding with the maximum impairment of corrosion resistance was not the same for each steel. (The maximum impairment of corrosion resistance did not normally occur at 650° C which is the usually

* A typical specification for the 18/8/Ti class of material is:

Carbon	0·15% maximum
Silicon	1·2% maximum
Manganese	2·0% maximum
Sulphur	0·05% maximum
Phosphorus	0·05% maximum
Nickel	7·5% minimum
Chromium	17·5% minimum
Titanium	Not less than 4 times the carbon content and not more than 1·0%

Fig. 3—Other half of the flanged pipe shown in Fig. 2. Complete penetration by wastage of the pipe wall had occurred at the zone of impaired corrosion resistance behind the back fillet weld attaching the flange to the pipe Actual size

Table I

CHEMICAL COMPOSITIONS OF 18/8/Ti FORGED BARS

Steel	C, %	Si, %	Mn, %	Ni, %	Cr, %	Ti, %	N, %
HF.1	0·11	0·53	0·59	9·48	18·39	0·45	0·020
HF.2	0·10	0·87	0·58	9·43	17·83	0·50	0·026
HF.3	0·11	1·26	0·55	9·45	18·63	0·48	0·020
HF.4	0·10	0·53	0·55	9·58	18·41	0·63	0·026
HF.5	0·11	0·81	0·60	9·46	18·04	0·83	0·022
HF.6	0·10	1·24	0·52	9·56	18·22	0·76	0·013
HF.7	0·10	0·63	0·72	9·51	18·41	1·12	0·018
HF.8	0·09	0·82	0·48	9·39	18·37	0·94	0·020
HF.9	0·10	1·27	0·60	9·35	18·50	1·30	0·020
Typical specification for 18/8/Ti steel	0·15 Max.	1·2 Max.	2·0 Max.	7·5 Min.	17·5 Min.	4 × (C%) Min. and not greater than 1·0	

accepted optimum temperature for development of maximum susceptibility to the so-called ' weld decay ' phenomenon.)

(iii) The application of a heat-treatment at 880° C for 1–2 h effectively restored the corrosion resistance of the material which had been impaired by heating in the 600–750° C range.

The small amount of laboratory work* which had previously been carried out on the ' 1300° C ' effect had indicated that

(i) The degree of susceptibility was different in 18/8/Ti steels of slightly different composition

(ii) The application of a (post-weld) heat-treatment at 880° C for 1–2 h did not ensure immunity to subsequent fissure wastage in nitric acid.

Having regard to the anomalous behaviour exhibited by welded 18/8/Ti austenitic steels when exposed to corrosive media of the nitric acid type and since very little information had been published on the effect of welding on the corrosion resistance of these steels, it was decided to undertake an examination of the influence of chemical composition on the susceptibility of the 18/8/Ti type of steel to impairment of corrosion resistance resulting from heating to the temperatures attained in the production of fusion-welded joints. It was hoped that this work would indicate a particular chemical composition of 18/8/Ti steel which would exhibit a much lower susceptibility to both the ' high ' and ' low ' temperature effects than hitherto observed. It was considered particularly important to obtain the minimum susceptibility to the ' high ' temperature effect, since it had already been established that the application of a post-weld heat-treatment at 880° C would effectively eliminate the impairment of corrosion resistance caused by heating to temperatures of within 600–750° C.

The present paper is presented in three parts, Parts I and II dealing with the 600–750° C and 1300° C effects respectively, in 18/8/Ti steels containing about 0·10% C, whilst Part III deals with 18/8/Ti steels containing about 0·06% C.

Part I—EFFECT OF THERMAL TREATMENT AT 600–750° C

EXPERIMENTAL WORK

Range of Chemical Composition Investigated

From the results of earlier work by the authors it appeared that the type of impairment of corrosion resistance under review was possibly associated with the distribution of C between Ti and Cr, and since the variation in Cr content between the different steels examined was not very significant, it was thought that the effect of varying the *titanium* content should be studied. The previous work on laboratory-tested specimens, taken in conjunction with observations on plant items, also appeared to indicate the possibility that the *silicon* content of the 18/8/Ti austenitic steels might be a factor in the mechanism of the impairment phenomenon, and it was decided, therefore, to investigate these two variables—titanium and silicon—within the limits of chemical composition for these two elements laid down in the typical commercial specification quoted in the Introduction. Three different silicon contents, viz. 0·5%, 0·8%, and 1·2% were examined at three different titanium contents, viz. 0·5%, 0·8%, and 1·0%, at a fixed carbon level of 0·10%.

The reason for keeping the *carbon* content constant at 0·10% (apart from the need to limit the number of variables under examination) was that the greater bulk of 18/8/Ti material for which considerable operating experience was available to the authors had a carbon content in the range 0·10–0·13%.

The remaining major elements in the 18/8/Ti composition of austenitic chromium nickel steel, viz. Cr, Ni, and Mn, were substantially the same in all the steels.

Preparation of the Laboratory Melts

The various compositions were obtained by melting ' stock ' material consisting of a plain 18/8 chromium–nickel steel of low carbon and low silicon contents

*Since the work was started, two papers[1, 2] describing some of the effects of welding on 18/8/Ti steels have been published.

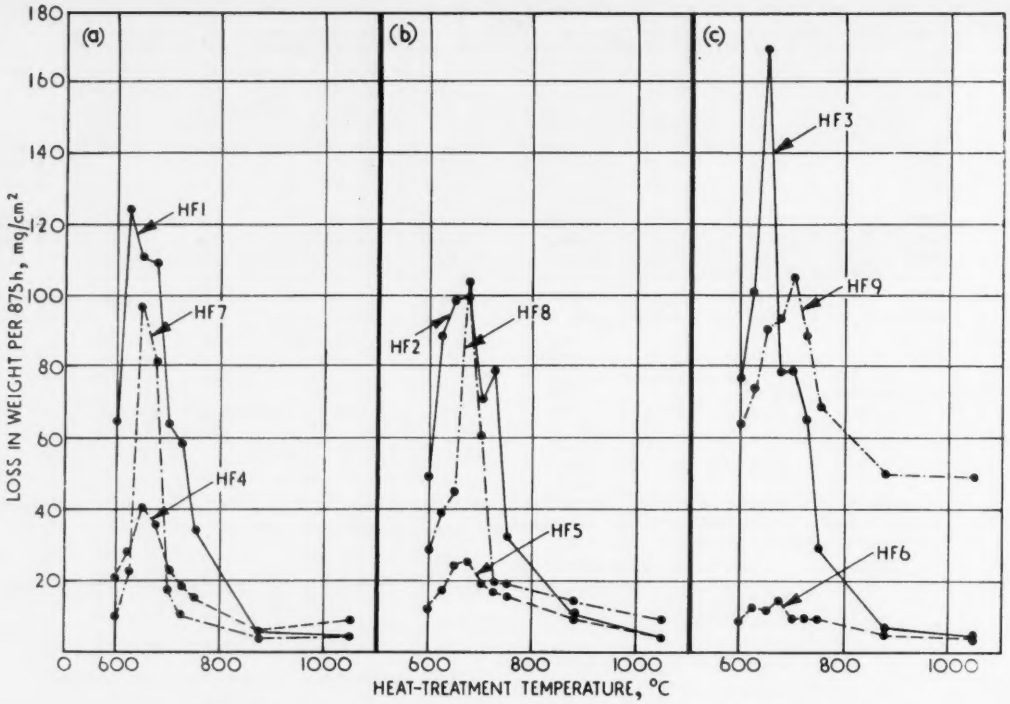

Fig. 4—Results of corrosion tests on 18/8/Ti forged bars containing 0·10% C

and adding the requisite amounts of Armco iron, manganese, chromium, nickel, and silicon as metals, and titanium in the form of ferro-titanium. The carbon content was adjusted to 0·10% by the addition of the appropriate amount of a non-alloy steel containing 1·3% C.

The materials were melted in a 15-lb capacity radio-frequency induction furnace, and the procedure during melting and casting was, as far as possible, 'standard' for each melt. The time of preparation of each melt was about 45 min, the completed melt being held at a steady temperature for 5 min before pouring. The melts were cast into a square vertical cast-iron mould 15 in. long, a preheated refractory feeder-head being used to obtain the maximum length of sound ingot. On solidification, the top (piped) portion and a 2-in. length at the bottom of the ingot were discarded.

Chemical analyses were carried out on portions taken from the top and bottom of the cropped ingot, in order to check the chemical composition and homogeneity of the material.

Thermal and Mechanical Treatment of the Ingots

After removing the top and bottom discards, the ingots—2¼in. square × about 8in. long—were examined for surface defects, which were eliminated by grinding and machining where necessary. The ingots were then heated slowly in an electric furnace to 1050° C, 'soaked' at that temperature for 30 min, and finally allowed to cool freely in air.

The ingots were then reheated to a temperature not exceeding 1200° C and forged to give flat bars 1½ in. wide × ¾ in. thick.

Chemical Compositions of the Forged Bars

Chemical analyses carried out on samples taken from the mid-length of each forged bar showed that the compositions were practically identical with those determined from the discarded portions of the ingot. The chemical compositions of the various bars are shown in Table I, from which it will be noted that, with the exceptions of steels *HF*.7 and *HF*.9, the compositions of the various bars conform with the limits quoted in the specification referred to in the Introduction. The steel *HF*.7 contains a titanium content of 1·12%, which is slightly higher than the upper limit of 1·0% usually specified, whilst steel *HF*.9 contains a still higher content of titanium (1·30%). Accurate control of the titanium content in small casts of 18/8/Ti steel is difficult to achieve owing to the high loss of titanium from the melt. The two steels *HF*.7 and *HF*.9 were included in the corrosion tests because it was felt that they would provide useful information on the characteristics of 18/8/Ti steels having a titanium content slightly greater than that normally supplied commercially.

Corrosion Tests on the Forged Bar Materials

Specimens for the corrosion tests were machined from the forged bars and given a heat-treatment at 1050° C for 15 min followed by cooling in air. The individual specimens were then heat-treated for 2 h at the appropriate temperature between 880° and 600° C.

Table II

EFFECT OF HEAT-TREATMENT AT 880° C FOR 2 h ON 18/8/Ti FORGED BAR MATERIALS OF IMPAIRED CORROSION RESISTANCE

Steel	C, %	Si, %	Ti, %	Loss in Weight in mg/cm² after 875 h in 30% Nitric Acid at 100° C, after Heat-Treatment at			
				1050° C (Fully Softened)	1050° + X° (Maximum Impairment of Corrosion Resistance)	1050° + X° + 880°	1050° + 880°
(1)	(2)	(3)	(4)	(5)	(6)	(7)	(8)
HF.1	0·11	0·53	0·45	5·5	122·7	11·3	5·8
HF.2	0·10	0·87	0·50	6·8	100·0	22·8	11·7
HF.3	0·11	1·26	0·48	5·7	170·6	14·6	8·0
HF.4	0·10	0·53	0·63	9·6	41·3	5·1	5·0
HF.5	0·11	0·81	0·83	8·0	26·6	10·8	10·7
HF.6	0·10	1·24	0·76	6·2	16·3	8·9	5·9
HF.7	0·10	0·63	1·12	5·1	97·4	7·7	3·5
HF.8	0·09	0·82	0·94	12·3	105·2	20·0	13·9
HF.9	0·10	1·27	1·30	50·9	107·5	86·7	52·5

The final preparation of the specimens before corrosion testing consisted of filing followed by rubbing on abrasive cloth, to make specimens approximately $1\frac{1}{8}$ in. long × $\frac{5}{8}$ in. wide × $\frac{1}{4}$ in. thick.

The specimens were immersed in the test medium contained in glass flasks by hanging them on the end of glass rods, which were in turn suspended from the tops of glass-air condensers, only one specimen being contained in a flask. The volume of test liquor contained in each flask was 250 cm³, the surface area of the test specimen in each case being approximately 15 cm². The various flasks were immersed in a water bath maintained at 100° C.

The corrosive medium used was an aqueous solution of nitric acid containing 30 wt.-% of nitric acid at 97–100° C, this solution having been previously found to be a useful 'sorting' medium for assessing the relative corrosion resistances of different compositions of 18/8/Ti steel. The specimens were exposed to the acid solution for a total period of 875 h, the specimens being weighed and the acid medium renewed at intervals of about 100, 150, 200, 200, and 200 h.

RESULTS OF THE CORROSION TESTS

Impairment of Corrosion Resistance Caused by Heating at 600–750° C

The results of the loss-in-weight corrosion tests are summarized in Figs. 4*a*, *b*, and *c*, in which the loss-in-weight over 875 h, expressed as mg/cm², has been plotted against the temperature of heat-treatment for the respective specimens. Steels of similar Si content have been plotted on the same chart, e.g. *a* in Fig. 4 refers to the steels of *lowest* silicon content. The relative susceptibilities of the nine steels to impairment of corrosion resistance by heating to a temperature in the 600–750° C range can be seen from Fig. 4 and it will be observed that all nine alloys showed impairment of corrosion resistance to a greater or less degree on heating in this range of temperature. In each of the three groups of alloys of similar Si content, the minimum degree of impairment of corrosion resistance occurs in the alloys of 'medium' Ti content, i.e. alloys *HF*.4, *HF*.5, and *HF*.6.

The remaining alloys, i.e. those of 'low' and 'high' Ti content, exhibit a similar degree of impairment amongst themselves, with the exception of alloy *HF*.3, containing a 'low' Ti and 'high' Si content, in which the greatest impairment occurred.

A further feature of note is that the 'intrinsic' corrosion resistance, i.e. that of the steels in the fully softened condition of eight of the alloys, is very similar but that steel *HF*.9, which has the high Ti content of 1·30% and the high Si content of 1·27%, is markedly inferior to the others. This indicates that the presence of titanium and silicon contents at the upper end of the usual specification limits for 18/8/Ti steels is undesirable, since the combination of high titanium and high silicon has an adverse effect on the general corrosion resistance of the material.

Effect of a Heat-Treatment at 880° C on 18/8 Ti Forged Bar Materials Previously Heated at 600–750° C

Mention has already been made of the effect of a heat-treatment at 880° C in eliminating almost completely the ill-effects which result from heating 18/8/Ti material at 600–750° C but in view of the wide differences exhibited by the nine 18/8/Ti forged bars in their susceptibility to impairment of corrosion resistance caused by heating at temperatures between 600–750° C, some further tests were carried out to ascertain whether the heat-treatment at 880° C was, in fact, beneficial over the range of chemical compositions covered by the laboratory-melted steels.

A specimen from each of the nine forged bars in the 'fully softened' condition was heat-treated for 2 h at that particular temperature in the 600–750° C range at which the steel had suffered the maximum impairment of corrosion resistance (*see* Figs. 4*a*, *b*, and *c*), after which it was heat-treated for 2 h at 880° C. The appropriate preparation of the surfaces of the specimens was carried out and they were subjected to the 'standard' corrosion testing procedure described above.

The effect of the 880° C heat-treatment on the various specimens is shown in Table II, and comparison of columns 5, 6, and 7 will show the effectiveness of the 880° C heat-treatment in almost completely eliminating the deleterious effect of heating at 600–750° C. (The degree of restoration appears to depend on the extent of the initial impairment, in general being greater where the latter is smaller.)

The final column (8) in Table II is included to show the effect of the application of the 880° C heat-treatment on material in the 'fully softened' condition, a feature of some practical significance in connection with welded plant items in 18/8/Ti steel which are almost invariably fabricated from material supplied in this condition. The heat-treatment has no detrimental effect, i.e. the corrosion resistance of the steels used in these tests is not impaired by the application of a final heat-treatment at 880° C. A heat-treatment at about 700° C, sometimes recommended for stress-relieving welded 18/8/Ti plant items, can be markedly detrimental to the corrosion resistance of many of the 18/8/Ti steels, as will be noted from column 6 of Table II.

The salient features revealed by the work carried out to determine the susceptibilities of the nine 0·10% carbon laboratory-prepared steels to impairment of corrosion resistance due to heating in the 600–750° C range of temperature are as follows:

Influence of Titanium Content

Quantitative assessment of the influence of Ti and Si is to some extent difficult because of the slight 'scatter' in the groups containing 'constant' Ti and in particular, those groups containing 'medium' and 'high' Ti, but very considerable differences in behaviour are exhibited by steels which conform with the requirements of a typical specification for 18/8/Ti material. Furthermore, there can be little doubt that the actual Ti content has a most important influence on the degree of susceptibility of the steel to impairment of corrosion resistance resulting from heating at 600–750° C.

It has been considered that the impairment phenomenon involves the co-behaviour of the Ti, Cr, and C present in the steel, a tentative explanation which has been put forward being as follows:

On applying the customary softening treatment to 18/8/Ti steel—i.e. heating at 1050° C—a certain proportion of the titanium carbide in the steel redissolves until equilibrium conditions are reached. On subsequent cooling in air, which is the commercial practice in the case of wrought products in this material, most of the redissolved carbides are retained in solution owing to the relatively rapid cooling from 1050° C. When the 'fully softened' material receives a subsequent heating at 600–750° C (such as occurs during a welding operation) both chromium carbide and titanium carbide are precipitated. It is postulated that this may take place because of the greater availability of Cr in the austenitic solid solution compared with that of Ti and that local depletion of the solid solution of Cr reduces the corrosion resistance of the material at the impoverished zones, primarily at the grain boundaries. The alloy is thus susceptible to preferential corrosion at these regions, the ultimate mechanism of the 600–750° C impairment being similar to that postulated in the better known and so-called 'weld decay' phenomenon which occurs in the unstabilized 18/8 austenitic Cr–Ni steels.

Table III

TITANIUM/CARBON RATIOS IN FORGED 18/8/Ti BARS

Steel	Total Ti, %	N, %	Ti, % (as Nitride)	Available Titanium, %	C, %	Actual Ti/C Ratio
HF.1	0·45	0·020	0·07	0·38	0·11	3·5 : 1
HF.2	0·50	0·026	0·09	0·41	0·10	4·1 : 1
HF.3	0·48	0·020	0·07	0·41	0·11	3·7 : 1
HF.4	0·63	0·026	0·09	0·54	0·10	5·4 : 1
HF.5	0·83	0·022	0·08	0·75	0·11	6·8 : 1
HF.6	0·76	0·013	0·04	0·72	0·10	7·2 : 1
HF.7	1·12	0·018	0·06	1·06	0·10	10·6 : 1
HF.8	0·94	0·020	0·07	0·87	0·09	9·6 : 1
HF.9	1·30	0·020	0·07	1·23	0·10	12·3 : 1

If this explanation of the 'low' temperature impairment phenomenon is substantially correct, the amount of titanium in the steel actually available for combination with carbon must be of considerable importance. The usual specification for Ti content in the 18/8/Ti class of material in which the C content of the steel does not exceed 0·15% is that the ratio of *total* Ti to C should not be less than 4 : 1. Theoretically, provided that all the Ti in the steel was available for combination with C, complete stabilization should be achieved by using this minimum Ti/C ratio, but, in fact, such a premise is not necessarily justified since some Ti may be combined with the element nitrogen, which is almost invariably present in small amounts in these steels. (A small quantity of the Ti may also be present in the form of non-metallic matter.) It is known from previous published work[1] that in the 18/8/Ti class of steel, practically all the nitrogen present is combined with Ti in the form of nitride (TiN) and consequently the whole of the Ti in the steel is not available for combination with C.

The amounts of nitrogen present in the nine experimental steels have been taken into account in compiling Table III; most of the alloys possess Ti/C ratios substantially greater than the 4 : 1 minimum specified. The ratios for steels *HF*.1 and *HF*.3 are, however, lower, being 3·5 : 1 and 3·7 : 1 respectively.

Nevertheless, despite the fact that in seven cases the Ti/C ratio exceeds the value of 4 : 1, none of the steels is immune from the 'low' temperature impairment phenomenon and it would therefore appear that titanium is not a complete stabilizing addition for the 18/8/Ti steels. The degree of susceptibility to impairment of corrosion resistance does not vary, however, in direct proportion with the Ti content, i.e. other things being equal, with the Ti/C ratio, for as will be seen from Fig. 4, in each group of three alloys of similar silicon content, the least impairment of corrosion resistance resulting from a heat-treatment at 600–750° C is exhibited by the steel having a Ti/C

ratio intermediate between the extremes. For example, of the three steels of medium Si content (Fig. 4*b*), alloy *HF*.2, having a Ti/C ratio of 4·1 : 1 (i.e. containing sufficient available titanium to combine with all the carbon in the steel), and alloy *HF*.8, having a Ti/C ratio of 9·6 : 1 (i.e. over twice the theoretical amount of Ti required for combination with the carbon), both exhibit corrosion rates of approximately 100 mg/cm² per 875 h, whilst alloy *HF*.5, containing the intermediate Ti/C ratio of 6·8 : 1, has the considerably lower corrosion rate of about 25 mg/cm² per 875 h.

The data in Fig. 4*b* also indicate that a high Ti/C ratio of 9·6 : 1 is not superior to the lower ratio of 4·1 : 1 as regards the 600–750° C impairment phenomenon in an alloy of 'medium' Si content, an observation which also appears to apply in the case of the steels of 'low' Si content.

The results of the corrosion tests thus suggest that for 18/8/Ti austenitic Cr–Ni steels containing about 0·10% C, immunity to impairment of corrosion resistance (i.e. the 600–750° C effect) cannot be obtained by ensuring that the steel contains an amount of Ti equal to or greater than that necessary to combine with the C in the steel. It is also indicated that, for 18/8/Ti steels containing 0·10% C, the minimum degree of impairment of corrosion resistance resulting from heating in the 600–750° C range of temperature is exhibited by those steels having a *nominal* Ti/C ratio of about 7·5 : 1, those of lower and higher ratios being inferior as regards susceptibility to the low-temperature impairment phenomenon.

The behaviour of the 18/8/Ti steels examined is not therefore fully in accordance with the tentative explanation of the 'impairment' phenomenon outlined earlier in this paper, since on that basis, the greater the titanium content of the steel, the less chance there should be of C precipitating as chromium carbide in preference to titanium carbide. It seems likely, therefore, that the mechanism of the impairment phenomenon is more complex than that originally envisaged.

Influence of Silicon Content

The influence of the Ti present in the steel is obviously so great that the effect of variation in the silicon content is difficult to assess from an examination of the data on the nine 18/8/Ti steels and although in the alloys containing 'medium' Ti there is a decrease in the degree of impairment as the Si content increases (*HF*.4, *HF*.5, and *HF*.6), it will also be noticed that in these three alloys the true Ti/C ratios are 5·4 : 1, 6·8 : 1, and 7·2 : 1 respectively. It is consequently much more likely that the progressive decrease in degree of impairment exhibited by these three steels is due to the Ti/C ratio approaching the optimum rather than to the increase in Si content.

Influence of Titanium and Silicon on the Intrinsic Corrosion Resistance

In addition to the observations on the degree of impairment of corrosion resistance exhibited by the nine *HF*. steels after heating to temperatures within the 600–750° C range, the corrosion tests also provided some information on the influence of Ti and Si on the 'intrinsic' corrosion resistance of the steels, i.e. the corrosion resistance of the material in the 'fully softened' condition, as generally supplied by the manufacturers.

Thus the values included in Table II for the 1050° C heat-treatment show that for alloys containing 'low' and 'medium' amounts of Ti there is no great difference in intrinsic corrosion resistance of the materials over the range of Si content which was examined. On the other hand, it would appear that for alloys containing 'high' amounts of Ti, an increase in Si content brings about a decrease in the intrinsic corrosion resistance (*HF*.7, *HF*.8, and *HF*.9). That this decrease in intrinsic corrosion resistance is not due to the influence of Si, *per se*, is evident from an examination of Fig. 4*c* which shows that whilst steels *HF*.3 ('low' Ti) and *HF*.6 ('medium' Ti) contain a 'high' Si content and possess an intrinsic corrosion resistance which may be regarded as normal, that of steel *HF*.9 containing both a 'high' Ti and a 'high' Si content is markedly inferior.

CONCLUSIONS

For 18/8/Ti steels containing about 0·10% C, the conclusions to be drawn from the work described in the foregoing are:

(i) The steels are susceptible to impairment of resistance to corrosion (by liquors of 'nitric acid' type) as a consequence of having been heated to temperatures in the 600–750° C range

(ii) 18/8/Ti steels may differ considerably in the degree of susceptibility to impairment of corrosion resistance, relatively minor variations in titanium content having a marked influence on the degree of impairment

(iii) The optimum nominal Ti/C ratio to give the least susceptibility to impairment of corrosion resistance appeared to be about 7·5 : 1, for the particular steels examined, this ratio being considerably higher than the 4 : 1 ratio generally accepted as the minimum value required for stabilization

(iv) The effect of the Si content on the degree of impairment of corrosion resistance is not significant compared with that of Ti

(v) The corrosion resistance of 18/8/Ti material which has been impaired as a result of heating in the 600–750° C range can be restored almost completely by the application of a heat-treatment at 880° C for 2 h

(vi) A steel having both a high Ti and a high Si content exhibits an intrinsic corrosion resistance which is markedly inferior to that of the remaining eight steels.

It is appreciated that these conclusions are based on corrosion tests carried out on experimental melts of steel which may not have been as homogeneous, or received as much hot-working, as commercial

Table IV

EFFECT OF HEATING TO 1300° C ON THE CORROSION RESISTANCE OF 18/8/Ti STEELS *A* AND *B*

Steel	Loss in Weight in mg/cm² after 875 h in 30% Nitric Acid at 100° C following Heat-Treatment at	
	1300° C (3–4 min)	1050° C (Fully Softened)
A	155·3	3·3
B	14·4	6·0

wrought 18/8/Ti steels of similar chemical composition. A comparison between the laboratory melts and some commercial wrought steels previously examined by the authors has, however, indicated that, as regards the phenomenon of impairment of corrosion resistance due to heating to temperatures in the 600–750° C range, the conclusions drawn from the work on the experimental forged bars can be regarded as a reliable indication of the behaviour of 18/8/Ti wrought commercial steels of comparable chemical composition.

Part II—EFFECT OF THERMAL TREATMENT AT 1300° C

The characteristic ' fissure ' which is formed in the parent material immediately adjacent to welds in some welded 18/8/Ti plant items exposed to onerous corrosive conditions of the nitric acid type is clearly associated with the particular thermal events which occur in such regions during the welding operation, and it can be shown that the more rapid local wastage which takes place in these zones is due to the fact that the parent material has been rendered less resistant to corrosion by the ' nitric acid ' type of medium as a consequence of having been heated to temperatures approaching the melting point of the material.

The quantitative investigation of this phenomenon by means of welded specimens presents considerable difficulty because welding usually involves steep temperature gradients in the plate material adjacent to the fusion zone. The temperatures attained are not susceptible to easy measurement or control for quantitative determination of the degree of impairment of corrosion resistance produced.

Since it was considered essential to make a quantitative assessment of the magnitude of any effects involved, it was necessary to employ thermal treatments which were known with a suitable degree of accuracy. It was therefore decided that the specimens which would be used in the corrosion tests should each have been uniformly heated to a predetermined temperature representative of the high temperatures produced in the ' heat-affected ' zone of the parent plate in which fissure formation occurs. The validity of this approach was checked by means of an initial experiment whereby specimens which had been welded (and therefore heated in a non-uniform manner) were compared with specimens of the same steels which had been heated uniformly to about 1300° C. The steels selected for this preliminary test were two commercial wrought 18/8/Ti steels (*A* and *B*) which had previously been examined by the authors and which were of the following chemical compositions:

	C, %	Si, %	Mn, %	Ni, %	Cr, %	Ti, %	N_2, %
Steel *A*	0·12	0·81	0·46	8·46	18·99	0·75	0·005
Steel *B*	0·08	0·52	0·49	7·81	17·82	0·32	0·005

Comparison of ' As Welded ' and ' Uniformly Heated ' specimens of steels A and B

The welded specimens were prepared by depositing one or more runs of weld metal on plate material which had previously been softened by heating at 1050° C. The specimens to represent the uniformly heated materials were prepared from the same plate (in the softened condition) and were then further treated at 1300° C for 3–4 min in a Globar electric furnace, followed by cooling in air.

Corrosion test specimens of the usual dimensions were prepared as described previously and immersed in 30% nitric acid at 100° C. They were taken out at intervals of 100–200 h and weighed, fresh acid being provided at each stage of the test.

The loss-in-weight values determined from the uniformly heated specimens of the steels *A* and *B* showed that the application of a heat-treatment for a few minutes at 1300° C to the steel *A* brought about a 50-fold impairment of corrosion resistance compared with that of the steel in the fully softened condition, whereas the steel *B* exhibited a very small impairment of corrosion resistance, the rate of wastage being approximately doubled as a consequence of the application of the 1300° C heat-treatment. Relevant values are shown in Table IV, in which is included for comparison the data for the same materials in the ' fully softened ' condition, i.e. cooled in air from 1050° C.

The significance of the different susceptibilities of steels *A* and *B* will be observed from the sections of welded corrosion test specimens of these steels (Fig. 5): specimen *A* suffered severe fissure attack, whereas steel *B* was almost completely unaffected. The difference between the two steels is, however, greater than would appear from the illustrations, since whilst steel *A* had been exposed for 3500 h to the test liquor (60% nitric acid at 100° C), the welded

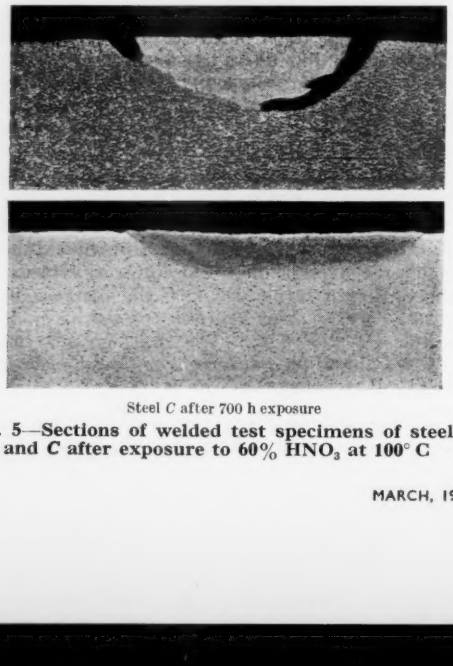

Fig. 5—Sections of welded test specimens of steels *A* and *C* after exposure to 60% HNO_3 at 100° C × 7

specimen of steel *B* shown in Fig. 5 had been exposed for a period of 7000 h in the same medium.

It was evident from the foregoing that the corrosion tests carried out on specimens heated uniformly to a suitably high temperature, such as 1300° C, provided a reliable indication of the relative susceptibilities of 18/8/Ti steels to fissure formation. The work also demonstrated that two representative 18/8/Ti steels could differ to a considerable degree in their susceptibility to the impairment of corrosion resistance caused by heating to 1300° C.

Effect of Heat-Treatment at 1300° C on Laboratory Melted Steels

A portion of each of the nine experimental steels referred to in Part I, initially in the ' fully softened ' condition, was heat-treated in a Globar electric furnace at 1300° C for 5 min, and then allowed to cool in air. After the usual preparation of the specimens, corrosion testing was carried out in 30% nitric acid at 100° C for 875 h, the progress of corrosion being followed at intervals by weighing the various specimens.

The results obtained from the nine steels (*HF*.1 to *HF*.9 inclusive) are given in Table V, the degree of impairment of corrosion resistance resulting from the application of the 1300° C heat-treatment being assessed from a comparison of the values given in columns *a* and *b*, which indicate respectively the effect of the 1300° C heat-treatment, and the intrinsic corrosion resistance of the steels (i.e. when cooled in air from 1050° C).

The values given in Table V show that the experimental work has revealed that the nine steels exhibited marked differences in their susceptibility to impairment of corrosion resistance after heating to 1300° C. Thus, steel *HF*.1 appeared to be immune from the ' high-temperature ' impairment of corrosion resistance, whilst steels *HF*.2, *HF*.3, and *HF*.4 were only slightly inferior. Steels *HF*.5 to *HF*.9 suffered severe impairment of corrosion resistance.

It is significant that the four steels showing little or no susceptibility to impairment as a result of the heat-treatment at 1300° C contain relatively low percentages of titanium (0·45–0·63%) with true Ti/C ratios between 3·5 : 1 and 5·4 : 1 and that the steels which exhibited severe impairment are those in which the titanium content is 0·75–1·30% with Ti/C ratios lying between 6·8 : 1 and 12·3 : 1. Evidently, in the range of steels examined, the titanium content is an important factor as regards the susceptibility to impairment of corrosion resistance brought about by a ' high-temperature ' heat-treatment.

The evaluation of the effect of silicon on the susceptibility of the 18/8/Ti steels to the 1300° C impairment phenomenon is difficult because, as has already been shown, the titanium content has a most pronounced influence and may completely mask the effect of the silicon. Nevertheless, it will be noted in Table V that for steels *HF*.1 and *HF*.3, in which there is only a small variation in titanium content between 0·45% and 0·48%, there is relatively little difference in the degree of impairment of corrosion resistance for a variation in silicon content between 0·53% and 1·26%. On the whole, the results of the experimental work suggest that the influence of silicon is not significant compared with that of titanium in regard to the 1300° C effect.

Table V

EFFECT OF HEAT-TREATMENT ON CORROSION RESISTANCE

Steel	Chemical Composition, %				Loss in Weight in mg/cm² after 875 h in 30% Nitric Acid at 100° C following Heat-Treatment at	
	C	Ti	Si	Ratio* Ti/C	(*a*) 1300° C	(*b*) 1050° C
HF.1	0·11	0·45	0·53	3·5 : 1	5·9	5·5
HF.2	0·10	0·50	0·87	4·1 : 1	16·8	5·8
HF.3	0·11	0·48	1·26	3·7 : 1	14·7	5·7
HF.4	0·10	0·63	0·53	5·4 : 1	15·0	9·6
HF.5	0·11	0·83	0·81	6·8 : 1	335·0	8·0
HF.6	0·10	0·75	1·24	7·2 : 1	207·0	6·2
HF.7	0·10	1·12	0·63	10·6 : 1	200·0	5·1
HF.8	0·09	0·94	0·82	9·6 : 1	300·0	12·3
HF.9	0·10	1·30	1·27	12·3 : 1	326·0	51·0

*Allowance has been made for titanium combined with nitrogen

From the series of corrosion tests which are summarized in Table V, it is evident that, for 18/8/Ti steels containing about 0·10% carbon, the presence of titanium in amounts greater than about 0·65–0·70% results in the steel being very susceptible to impairment of corrosion resistance on heating to 1300° C. Welds in steels of such chemical compositions would therefore be expected to show considerable fissure wastage of the parent material at the welded joint when exposed to corrosive conditions of the nitric acid type. In fact laboratory corrosion tests in which welded specimens of steels *HF*.1–*HF*.9 were exposed to boiling 60% nitric acid (a severely corrosive reagent) have demonstrated the expected results that steels *HF*.1–*HF*.4 showed little fissure attack, whilst the latter was appreciable in steels *HF*.5–*HF*.9.

Effect of Subsequent Heat-Treatments on 18/8/Ti Material Heated to 1300° C

Heat-Treatment at 880° C for 2 h

In Part I, it was shown that when a heat-treatment for 2 h at 880° C was applied to 18/8/Ti austenitic Cr–Ni steels of impaired corrosion resistance (following heat-treatment at 600–750° C), a considerable improvement of corrosion resistance resulted. A heat-treatment at 880° C was therefore applied to material which had previously been treated to 1300° C and the effect investigated by means of corrosion tests which were carried out under the conditions described earlier, viz. exposure to 30% nitric acid at 100° C for 875 h.

It was found from the corrosion tests, particulars of which are given in Table VI, that the application of a heat-treatment for 2 h at 880° C on the 18/8/Ti forged bar materials previously heated to 1300° C

Table VI

EFFECT OF A HEAT-TREATMENT AT 880° C FOR 2 h FOLLOWING AN INITIAL HEAT-TREATMENT AT 1300° C

Steel	Chemical Composition				Loss in Weight in mg/cm² after 875 h in 30% Nitric Acid at 100° C following Heat-Treatment at	
	C, %	Ti, %	Si, %	Ratio* Ti/C	1300° C (*a*)	1300° + 880° C (*b*)
HF.1	0·11	0·45	0·53	3·5 : 1	5·9	47
HF.2	0·10	0·50	0·87	4·1 : 1	16·8	207
HF.3	0·11	0·48	1·26	3·7 : 1	14·7	309
HF.4	0·10	0·63	0·53	5·4 : 1	15·0	200
HF.5	0·11	0·83	0·81	6·8 : 1	335·0	388
HF.6	0·10	0·75	1·24	7·2 : 1	207·0	182
HF.7	0·10	1·12	0·63	10·6 : 1	200·0	52
HF.8	0·09	0·94	0·82	9·6 : 1	300·0	150
HF.9	0·10	1·30	1·27	12·3 : 1	326·0	128

* Allowance has been made for titanium combined with nitrogen

had rather different effects on the individual steels, and that in most cases deterioration of corrosion resistance occurred, whilst in some cases a slight improvement resulted. In every case, however, the corrosion resistance of the material which had been heated at 880° C following heat-treatment at 1300° C was very much inferior to the 'intrinsic' corrosion resistance, as will be observed from a comparison between column *b* of Table VI and column *b* of Table V.

The results of the corrosion tests shown in Table VI indicate that a heat-treatment at 880° C following a heat-treatment at 1300° C (*a*) brought about a marked reduction in the corrosion resistance of steels *HF*.1–*HF*.4, previously shown to be little affected by the 1300° C treatment alone, (*b*) had relatively little effect on steels *HF*.5 and *HF*.6, and (*c*) improved the corrosion resistances of steels *HF*.7, *HF*.8, and *HF*.9.

Thus, the application of a subsequent heat-treatment at 880° C for 2 h to material which had previously been heated for a short time at 1300° C appears to be deleterious in those alloys containing about 0·10% carbon and having a Ti/C ratio less than about 5·5 : 1. On the other hand, in alloys having Ti/C ratios greater than about 7 : 1, the 880° C treatment may have a beneficial effect in partially removing the ill-effects of the 1300° C heat-treatment.

It is apparent, therefore, that the effect of a post-weld heat-treatment at 880° C on 18/8/Ti steels which have been welded will depend upon the titanium content. The heat-treatment will not, in general, reduce the vulnerability to fissure wastage and in some compositions may actually render the steel much more susceptible.

Heat-Treatment at 1050° C for 10 min

In view of the varying effects of heat-treatment at 880° C on the steels which had suffered impairment of corrosion resistance as a result of prior heating to 1300° C, the effect of applying a heat-treatment at 1050° C for 10 min was examined in the case of three of the forged 18/8/Ti bar materials, *HF*.4, *HF*.7, and *HF*.8.

Table VII

EFFECT OF HEAT-TREATMENT AT 1050° C AFTER INITIAL TREATMENT AT 1300° C ON STEELS *HF*.4, *HF*.7, AND *HF*.8

Steel	Chemical Composition				Loss in Weight in mg/cm² after 875 h in 30% Nitric Acid at 100° C following Heat-Treatment at	
	C, %	Ti, %	Si, %	Ratio* Ti/C	1300° C	1300° + 1050° C
HF.4	0·10	0·63	0·53	5·4 : 1	15·0	10·7
HF.7	0·10	1·12	0·63	10·6 : 1	200·0	30·7
HF.8	0·09	0·94	0·82	9·6 : 1	300·0	29·0

* Allowance has been made for titanium combined with nitrogen

It was found in each case that a 10-min heat-treatment at 1050° C was beneficial in that a considerable degree of restoration of corrosion resistance was achieved, as will be seen from Table VII. The effect produced is very different from that produced by heat-treatment at 880° C for 2 h and it is evident that, in the case of welded items in 18/8/Ti steel within the range of chemical composition included in the experimental work, the only heat-treatment which could, with certainty, reduce the susceptibility to fissure wastage would involve heating to 1050° C, a heat-treatment which for most welded plant items would be quite impracticable.

DISCUSSION

The work described indicated without doubt that titanium-bearing 18/8/Ti steels containing about 0·10% carbon, and whose chemical compositions lie in the range covered by a typical specification for this class of austenitic steel, may suffer severe impairment of corrosion resistance to the nitric acid type of corrosive as a result of heating to about 1300° C. Such conditions arise during the welding operations and consequently fissures are likely to be formed at welds in these materials when they are exposed to onerous conditions of the nitric acid type.

The susceptibility of the 18/8/Ti steels to fissure wastage is greatly influenced by the amount of titanium present in the alloys, those steels containing smaller amounts (0·4–0·7%) of this element exhibiting relatively slight susceptibility, whilst the steels containing larger amounts of titanium (0·8–1·30%) are strongly susceptible to the 1300° C impairment of corrosion resistance.

Whilst the precise effect of silicon cannot be ascertained from the data available from the present series of tests, the trend of the results obtained suggests that within the limits of chemical composition investigated, silicon is not a significant factor as regards the '1300° C' impairment of corrosion resistance.

An important characteristic of the phenomenon of impairment of corrosion resistance which results from

Table VIII

CHEMICAL COMPOSITION OF FORGED BARS CONTAINING 0·06% C

Steel	Chemical Composition							Ratio of Ti/C*
	C, %	Si, %	Mn, %	Ni, %	Cr, %	Ti, %	N, %	
HF.12	0·06	0·51	0·64	9·52	19·08	0·41	0·031	5·1 : 1
(*HF*.4)	(0·10)	(0·53)	(0·55)	(9·58)	(18·41)	(0·63)	(0·026)	(5·4 : 1)
HF.13	0·06	0·86	0·59	9·64	19·00	0·25	0·033	2·3 : 1

* Allowance has been made for titanium combined with nitrogen

heating to 1300° C is that the deleterious effect of this thermal treatment cannot be removed by a subsequent heat-treatment at 880° C and that in certain compositions of steel the 1300° C impairment is grossly aggravated by a further heat-treatment at 880° C.

Thus, the application to a welded item of the heat-treatment at 880° C which is most beneficial in eliminating the impairment of corrosion resistance caused by heating at 600–750° C, will not eliminate the tendency to fissure formation at welds subject to onerous corrosive conditions. In fact, the available data indicate that this tendency cannot be eliminated by any practicable heat-treatment which can be applied to most welded plant items and in this respect the 1300° C effect is quite different from that brought about by heating in the 600–750° C range.

Part III—18/8/Ti STEELS CONTAINING 0·05–0·06% CARBON

Introduction

Since fissure wastage gives rise to a more serious ' stress-raiser ' than that produced as a result of the impairment of corrosion resistance caused by heating to the 600–750° C range of temperature, it was obviously desirable to attempt to find a composition of 18/8/Ti steel with the minimum susceptibility to the 1300° C effect. The steels of lower titanium content were apparently least susceptible to the fissure effect and accordingly it was decided to examine the behaviour of a steel having a somewhat lower titanium content than any of the other laboratory-melted steels studied. At the same time it was thought necessary to employ the optimum Ti/C ratio which had been indicated by the work described in Part I to give rise to the minimum susceptibility to the 600–750° C phenomenon, this ratio being about 7 : 1. The combination of a low titanium content and a Ti/C ratio of the desired order could not of course be obtained with a carbon content of 0·10% and it was thus necessary to make a steel having a lower carbon content, the value of 0·06% being chosen.

In view of the influence of the titanium content on the characteristics of the 18/8/Ti steels containing 0·10% C, a further steel containing 0·06% C but with a lower Ti/C ratio was prepared, in order to obtain some information on the effect of variation in titanium content at this lower carbon level.

The manufacture of the two ingots of the 0·06% C 18/8/Ti steels (*HF*.12 and *HF*.13) and the preparation of the forged bars was carried out as described in Part I. The chemical analyses determined on the bars forged from the ingots are shown in Table VIII, which also includes the composition of the 0·10% C steel *HF*.4 for purposes of comparison with steel *HF*.12.

Table VIII shows that the nominal Ti/C ratio of the 0·06% C steel *HF*.12 is almost 7 : 1 but when allowance is made for the amount of titanium combined with the nitrogen present in the steel, the actual revised Ti/C ratio is about 5 : 1, this being somewhat lower than that intended.

The other 0·06% C steel *HF*.13 has a nominal Ti/C ratio of 4·1 : 1, but when the nitrogen content of the steel is taken into account, the actual ratio is 2·3 : 1.

EXAMINATION OF TWO LABORATORY-PREPARED 0·06% C 18/8/Ti STEELS

Effect of Heat-Treatment at 1300° *C, and at* 880° *C on Material Previously Heated at* 1300° *C*

Portions of the forged bars—initially in the ' fully softened ' condition—were heat-treated for a few minutes at 1300° C and then allowed to cool in air. Parts of these heat-treated specimens were given a further heat-treatment at 880° C for 2 h, as in the 0·10% C steels referred to in Parts I and II.

The assessment of relative resistance to corrosion was made by means of corrosion-test specimens prepared from the heat-treated portions of the forged bars and immersed in 30% nitric acid solution at 100° C for 875 h, as described in Part I.

The effect on the corrosion resistance of the change in the carbon content from 0·10% to 0·06%, in conjunction with a lower titanium content giving approximately the same actual Ti/C ratio (5 : 1), is shown by a comparison of the corrosion data for steels *HF*.4 and *HF*.12 given in Table IX. It will be noted that the 0·10% C steel *HF*.4 shows slight susceptibility to impairment of corrosion resistance after heating at 1300° C but that the lower carbon steel *HF*.12 is completely unaffected (columns *a* and *c*). Furthermore, the application of a further heat-treatment at 880° C to steel *HF*.4 causes a severe reduction in corrosion resistance, whereas steel *HF*.12 is unaffected (columns *b* and *c*).

The low-carbon steel *HF*.12 is, in fact, superior to any of the nine experimental steels previously tested, as regards freedom from impairment of corrosion resistance resulting from heating to 1300° C and would be relatively immune from susceptibility to fissure wastage, either in the ' as welded ' or the ' welded and stress relieved (880° C) ' condition.

A further point of interest emerges from a comparison of these two steels (*HF*.4 and *HF*.12) in respect of the data in Table IX. The ' intrinsic '

Table IX

EFFECT OF HEAT-TREATMENT AT 1300° C AND AT 880° C ON MATERIAL WHICH HAD BEEN PREVIOUSLY HEATED AT 1300° C

Steel	Loss in Weight in mg/cm² after 875 h in 30% Nitric Acid at 100° C following Heat-Treatment at		
	1300° C (*a*)	1300° + 880° C (*b*)	1050° C (*c*)
HF.12 (0·06% C)	1·9	2·0	1·8
(*HF*.4) (0·10% C)	(15·0)	(200·0)	(9·6)
HF.13 (0·06% C)	2·0	10·0	2·2

corrosion resistance of the 0·06% C steel *HF*.12 is superior to that of steel *HF*.4 which contains 0·10% C. The rate of wastage due to corrosion by the nitric acid type of medium would, in fact, be less than that of any of the 0·10% C 18/8/Ti steels, a feature which is of considerable importance in relation to life of plant items under onerous corrosive conditions.

The influence of the titanium content at a carbon level of 0·06% will be observed from a comparison of the data in Table IX for *HF*.13 with that for *HF*.12. Thus, whilst the steel of very low titanium content is not inferior as regards intrinsic corrosion resistance or susceptibility to the 1300° C effect, it exhibits appreciable impairment of corrosion resistance when a heat-treatment at 880° C is applied, following an initial heat-treatment at 1300° C.

Effect of Heat-Treatment at 600–750° C

Portions of the 0·06% C steels *HF*.12 and *HF*.13, initially in the ' fully softened ' condition, were heat-treated for 2 h at various temperatures in the range 550–880° C as described in Part I of this paper. Corrosion specimens were tested for 875 h in the ' standard ' corrosive medium (30% nitric acid at 100° C), the progress of corrosion being followed in the usual manner.

The results of these tests are given in Fig. 6, from which it is evident that the 0·06% C steel *HF*.12 exhibits only slight impairment of corrosion resistance on heating at 600–750° C. The superiority of the lower carbon content for a given Ti/C ratio will be evident from a comparison of the corrosion/heat-treatment plot of the 0·06% carbon steel *HF*.12 with that of the 0·10% carbon 18/8/Ti steel *HF*.4, which has a similar Ti/C ratio to that of steel *HF*.12 (approximately 5 : 1) and otherwise differs significantly only in respect of the carbon content, which is 0·10%.

A comparison of the corrosion/heat-treatment plots of the two 0·06% C steels *HF*.12 and *HF*.13 shows, however, that the advantage of the lower carbon content as regards the 600–750° C effect is dependent upon the presence of a sufficiency of titanium, the steel with a true Ti/C ratio of 2·3 : 1 being appreciably susceptible to the effects of a heat-treatment at 600–750° C.

Thus, not only is a 0·06% carbon 18/8/Ti steel with a true Ti/C ratio of about 5 : 1 superior to the best of the nine 0·10% C 18/8/Ti laboratory-prepared steels in regard to susceptibility to the fissure effect, but it also exhibits appreciably less susceptibility to the 600–750° C impairment effect than any of the higher-carbon steels. In addition, the 0·06% C 18/8/Ti steel shows a very satisfactory ' intrinsic ' corrosion resistance. The presence of sufficient titanium to provide a Ti/C ratio of about 5 : 1 does, however, appear to be essential in the 0·06% C 18/8/Ti steel, since steel *HF*.13 which has a Ti/C ratio considerably lower than 5 : 1, whilst not being susceptible to the 1300° C effect, shows impairment of corrosion resistance on heating in the 600–750° C range.

EXAMINATION OF COMMERCIAL WROUGHT 18/8/Ti STEELS CONTAINING ABOUT 0·06% CARBON

Although the results of the corrosion tests on the laboratory-prepared experimental steel *HF*.12 containing 0·06% C had indicated that a lower carbon content was advantageous as regards low susceptibility of the steel to both the 600–750° C impairment and the 1300° C impairment, the work had been carried out on one steel only, which might have exhibited some abnormality despite the efforts made to avoid variables in the steelmaking and subsequent hot manipulation and heat-treatment processes.

It was therefore desirable to compare the results determined on this steel with those obtained on material which had been through the usual commercial manufacturing procedure. Such a comparison was carried out and in the course of this work nine different casts of steel were examined.

The various steels were obtained in the form of rolled plate material ranging between ¼ in. and ½ in. thick, of the chemical compositions indicated in Table X. Their carbon contents were in the range 0·05–0·06% and all conformed with the requirements of the typical specification quoted earlier.

Effect of Heat-Treatment at 1300° C, and at 880° C on Material Previously Heated at 1300° C

Material from the nine commercial steels was heat-treated and specimens prepared for corrosion testing as described and the ' standard ' corrosion test procedure was applied. The results are shown in Table XI.

Comparison of columns *a* and *c* of Table XI indicates that all the commercial steels containing 0·05–0·06% C examined showed similar characteristics to the laboratory-prepared steel *HF*.12 in that no significant impairment of corrosion resistance results from a heat-treatment at 1300–1350° C. Furthermore, from column *b* of Table XI it will be noted that, with

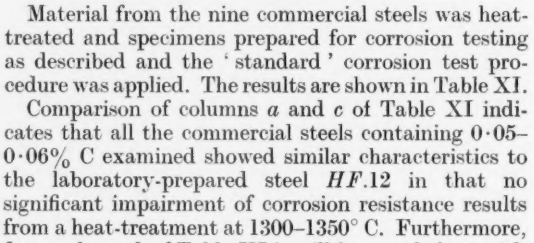

Fig. 6—Results of corrosion tests on steels *HF*.4, *HF*.12, and *HF*.13

Table X

CHEMICAL COMPOSITIONS OF COMMERCIAL 18/8/Ti STEELS CONTAINING 0·06% MAX. C

Steel Index No.	Composition							Ratio* Ti/C
	C, %	Si, %	Mn, %	Ni, %	Cr, %	Ti, %	N, %	
C	0·06	0·61	0·50	8·37	18·28	0·35	0·026	4·4 : 1*
D	0·06	0·65	0·57	8·18	17·65	0·35	0·011	5·1 : 1*
E	0·06	0·63	0·57	8·12	18·11	0·39	0·016	5·6 : 1*
F	0·05	0·49	0·50	8·13	18·13	0·37	0·018	6·6 : 1*
G	0·06	0·39	1·34	10·87	18·35	0·44	0·009	6·8 : 1*
H	0·06	0·85	0·50	8·03	17·52	0·49	0·017	7·2 : 1*
J	0·05	0·70	0·50	9·40	19·82	0·38	n.d.	7·6 : 1
K	0·06	0·65	0·58	8·29	18·26	0·52	0·014	7·7 : 1*
L	0·06	0·48	0·30	9·44	20·10	0·60	0·012	9·3 : 1*

* Allowance has been made for titanium combined with nitrogen

one exception, the application of a heat-treatment at 880° C following a heat-treatment at 1300–1350° C does not cause any further deterioration in corrosion resistance.

Table XI shows also that steel *H*, whilst being unaffected by a heat-treatment at 1300° C, shows slight susceptibility to impairment of corrosion resistance after further heat-treatment at 880° C. Whilst the 'out of line' behaviour of steel *H* might possibly be ascribed to the somewhat lower-than-average chromium content, steel *D*, which is only very slightly richer in chromium, does not behave in a similar fashion after double heat-treatment at 1300° and 880° C.

The results of these tests demonstrate that the 0·06% C experimental steel *HF*.12 is not abnormal in its response to the 'high-temperature' heat-treatment but is representative of commercial steels of similar composition. The work shows, in fact, that an 18/8/Ti steel of about 0·06% C exhibits much less susceptibility to the impairment of corrosion resistance which occurs when similar steels of 0·10% C are heated to 1300–1350° C. In addition, the greatly reduced susceptibility obtains over a wide range of Ti/C ratios, viz. 4·4 : 1 to 9·3 : 1.

In consequence, therefore, these lower-carbon steels might be expected to be much less liable to the fissure type of wastage at welded joints, whether or not the welds have been given a post-weld heat-treatment at 880° C. A further point to be noted from column *c* of Table XI is that the intrinsic corrosion resistance of all the 0·05–0·06% C steels are satisfactory.

Corrosion Tests on Welded Samples of 0·06% C 18/8/Ti Commercial Steels *G*, *J*, and *L*

In view of the fact that the work described above had indicated the virtual non-susceptibility of both the lower-carbon commercial steels and the lower-carbon experimental steel *HF*.12 to impairment of corrosion resistance as a result of heat-treatment at 1300/1350° C, some welded specimens of three of the commercial alloys, both with and without a subsequent heat-treatment at 880° C for 2 h, were prepared and subjected to corrosion tests to observe whether, in fact, there was any indication of fissure wastage at the welds. Since the various steels had been shown to be relatively immune from any deleterious effect of the 1350° C heat-treatment when exposed to 30% nitric acid at 100° C, the corrosion test applied was more severe than the 'standard' test, and in fact consisted of exposure to 60% nitric acid at 100° C for 4000 h.

The corrosion-test specimens were prepared from butt-welded plate, welded by the metal-arc process using the usual niobium-bearing 18/8 electrodes.

On conclusion of the corrosion tests the specimens were sectioned for microscopical examination in order to determine the severity of the fissure wastage which had taken place. Details of these observations are given in Table XII.

Thus, microscopical examination showed that despite the very onerous corrosive conditions employed in these tests, the amount of fissure attack which occurred in the welded specimens had been very slight and further, that the application of a heat-treatment at 880° C for 2 h had not increased the susceptibility to fissure formation.

Effect of 600–750° C. Thermal Treatment on Commercially Produced 0·06% C Steels

As it was also considered desirable to examine the susceptibility of these steels to the 600–750° C effect, portions of seven of the steels were given a preliminary softening treatment at 1050° C for 15 min, followed

Table XI

EFFECT ON COMMERCIAL STEELS OF HEAT-TREATMENT AT 1300° C AND AT 880° C ON MATERIAL PREVIOUSLY HEATED TO 1350° C

Steel	Chemical Composition, %			Loss in Weight in mg/cm² after 875 h in 30% Nitric Acid at 100° C following Heat-Treatment at		
	C	Ti	Ratio* Ti/C	1300° C (*a*)	1300° + 880° C (*b*)	1050° C (*c*)
C	0·06	0·35	4·4 : 1	2·0	2·6	1·7
D	0·06	0·35	5·0 : 1	3·2	3·8	2·5
E	0·06	0·39	5·6 : 1	2·4	3·8	2·0
F	0·05	0·37	6·6 : 1	3·2	3·4	2·0
G	0·06	0·44	6·8 : 1	0·9†	0·9‡	0·8
H	0·06	0·49	7·2 : 1	3·9	13·8	2·4
J	0·05	0·38	7·6 : 1	3·4†	1·8‡	1·3
K	0·06	0·52	7·7 : 1	3·5	4·8	2·2
L	0·06	0·60	9·3 : 1	1·5†	1·2‡	1·1

* Allowance has been made for titanium combined with nitrogen
† 1350° C ‡ 1350° C + 880° C

Table XII

FISSURE WASTAGE IN WELDED SPECIMENS OF WROUGHT COMMERCIAL 0·06% C STEELS AFTER 4000 h TEST IN 60% NITRIC ACID AT 100° C

Steel	Fissure Wastage	
	As Welded	Welded and Heat-Treated for 2 h at 880° C
G	Trace	Trace
J	Very slight	Trace
L	Very slight	Trace

by cooling in air, and individual specimens were subsequently heat-treated for 2 h at various temperatures in the range 550–880° C.

These specimens were then exposed to the 'standard' corrosion test conditions, i.e. 30% nitric acid solution at 100° C for 875 h, and the loss-in-weight data for the various steels determined. The results are shown in graphical form in Fig. 7, from which it will be observed that in each of the seven low-carbon commercial 18/8/Ti steels the degree of impairment of corrosion resistance on heating at 600–750° C was very small, being of the same order as that of the 0·06% carbon experimental steel *HF*.12 (*see* Fig. 6).

It is also of interest to note that despite the relatively wide range of Ti/C ratios covered by the seven steels, viz. 4·4 : 1 to 9·3 : 1, the degree of impairment of corrosion resistance is not significantly different and it would therefore appear that in the range of the 0·06% C steels studied, variation in titanium content between about 0·3 and 0·6% does not appreciably affect the degree of impairment of corrosion resistance produced by heating at 600–750° C. In this respect, therefore, the behaviour of the 18/8/Ti steels containing 0·05–0·06% carbon differs considerably from that of steels containing 0·10% carbon.

DISCUSSION

The experimental work on the 0·06% carbon 18/8/Ti laboratory prepared steel *HF*.12 has shown that, at this level of carbon content, an 18/8/Ti steel of the usual composition as regards chromium and nickel content is very much less susceptible to either the 1300° C or the 600–750° C impairment of corrosion resistance than the steels of higher (0·10%) carbon content but of otherwise similar chemical composition. The commercially produced 0·05–0·06% C 18/8/Ti steels also exhibit similar characteristics to the laboratory-prepared steel and show that the latter is in no way abnormal but can be regarded as accurately representative of 18/8/Ti steels of the chemical composition concerned.

Whilst steel *HF*.12 and the nine 0·05–0·06% C commercially produced steels indicate that minimum susceptibility to the 600–750° C effect occurs over a 'true' Ti/C range of about 4·5 : 1 to 9·5 : 1, the tests on the laboratory-prepared steel *HF*.13, with 0·06% carbon and a true Ti/C ratio of 2·3 : 1 show that there exists a lower limit for the titanium content, below which appreciable susceptibility to the 600–750° C effect is likely to be exhibited.

GENERAL DISCUSSION

The work described in this paper has shown that both titanium and carbon are very important factors in regard to the phenomenon of impairment of corrosion resistance (to corrosive media of the nitric acid type) exhibited by some 18/8/Ti steels after having been heated to the temperatures involved in the welding operation. The precise effect of silicon has not been ascertained but it appears to be relatively insignificant compared with that of C and Ti.

In the case of the 18/8/Ti steels containing 0·10–0·13% carbon, the titanium content is a dominant factor in the susceptibility to impairment of corrosion resistance, a 'low' titanium content being decidedly detrimental in regard to the 600–750° C effect, but associated with relatively slight susceptibility to the 1300° C effect. The presence of a slightly higher titanium content, as in the alloys containing 'medium' amounts of Ti, is associated with considerable susceptibility to the 1300° C effect but a lower susceptibility to the 600–750° C impairment, whilst a 'high' percentage of Ti appears to be quite detrimental in respect of both effects. It is evident that the role of the Ti is not the same in both types of impairment of corrosion resistance.

The fact that relatively small variations in the Ti content of the 0·10–0·13% C 18/8/Ti steels produce widely differing effects on the susceptibility to impairment of corrosion resistance of the forged bars provides a possible explanation for some apparent anomalies which have been observed in the past in welded plant items exhibiting either the 600–750° C. or the 1300° C effect. In the case of the vessel cited on page 242, small differences in chemical composition of the abutting plates at the welded seam could account for the fact that, of two plates concerned, one showed severe fissuring, whilst the other was unaffected.

The rather unexpected findings that (i) freedom from the 'low' temperature impairment cannot be achieved with a 0·10% C 18/8/Ti steel by providing a nominally adequate Ti/C ratio of 4 : 1, and (ii) for a given steel the Ti/C ratio associated with minimum susceptibility to the 'low' temperature effect was not the same as that required for minimum susceptibility to impairment resulting from heating to 'high' temperatures, suggests that the accepted explanation of the 'stabilizing' effect of titanium in 18/8/Ti austenitic steels is not adequate and that the Ti/C relationship is more complex than is usually envisaged.

Another interesting feature revealed by the work is that whilst the impairment of corrosion resistance which results from heating to 600–750° C is effectively eliminated by the application of a heat-treatment at 880° C, this heat-treatment does not in general eliminate the ill-effects resulting from heating to 1300° C. In fact, certain compositions of the 0·10% C 18/8/Ti steels are rendered more susceptible to fissure wastage by the application of the 880° C treatment. A heat-treatment at 1050° C is, however, effective in reducing considerably the deleterious effect of heating to 1300° C but the application of a heat-treatment at a temperature as high as 1050° C is rarely practicable for most welded plant items in 18/8/Ti steel.

The important influence of the carbon content is shown by the marked differences in response to thermal treatments between the 18/8/Ti steels which contain

0·06% C and those containing 0·10% C, the former being very much less susceptible than the latter in regard to susceptibility to both '600–750° C' and the '1300° C' effects. As a consequence, heating to the temperatures attained in the welding operation does not bring about any significant impairment of corrosion resistance (to the nitric acid liquors under examination) of the 0·06% 18/8/Ti material and the performance of welded plant items in the 0·06% C 18/8/Ti material under onerous corrosive conditions of the 'nitric acid' type is therefore likely to be greatly superior to that of the corresponding 18/8/Ti steels containing about 0·10% C.

Although the 0·06% C 18/8/Ti steels exhibit the minimum susceptibility to the 1300° C effect, the influence of the titanium content is still significant in that a steel with 'inadequate' titanium content is appreciably susceptible to the 600–750° C effect and it is evident that a nominal Ti/C ratio of 4 : 1 is not adequate to provide the minimum susceptibility to the 600–750° C effect.

A further feature of importance which has emerged is that the 'intrinsic' corrosion resistance of the steels containing about 0·06% carbon (i.e. the corrosion resistance of the material in the 'fully softened' condition, as commonly supplied) is, in general, somewhat superior to that of the 0·10% C 18/8/Ti steels of similar chromium content. The amount of chromium present in the steel is, of course, an important factor in determining the corrosion resistance of 18/8/Ti steels to 'nitric acid' conditions, but a comparison of the corrosion-test results for experimental steels *HF*.4 (0·10% C) and *HF*.12 (0·06% C) of almost identical chromium content, shows the superiority of the lower-carbon material. It is of interest to record that the intrinsic corrosion resistance of the 0·06% C commercial steels described in this paper was, in fact, better than any of the commercial 18/8/Ti steels containing about 0·10% C which have been examined by the authors, and since, in the absence of any local susceptibility to corrosion wastage, it is the 'intrinsic' corrosion resistance of the material which determines the rate of wastage of a particular plant item, the 0·06% C steel containing from about 0·35 to about 0·60% titanium is therefore somewhat superior in this respect to the 18/8/Ti steels containing 0·10% carbon.

It will thus be clear that for welded chemical plant items exposed to onerous conditions of the 'nitric acid' type the 0·06% C 18/8/Ti material, being relatively immune from the local impairment phenomena, is a superior material of construction to the 18/8/Ti steels having a content of 0·10% C and above.

Apart from the rather better intrinsic corrosion resistance of the 0·06% C material, it does not require to be heat-treated (at 880° C.) after welding in order to eliminate regions of impaired corrosion resistance and consequently a post-weld heat-treatment will not be required except in a limited number of instances, such as items exposed to the risk of stress-corrosion cracking. This is also an important consideration particularly where site welding is carried out for repair purposes or modification. (Where it is necessary for an item to be placed into service in a stress-free condition, a heat-treatment at 880° C can, of course, be applied without rendering the steel vulnerable to fissure wastage.)

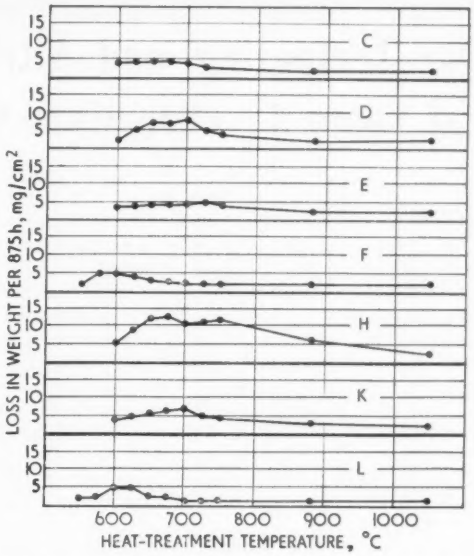

Fig. 7—Results of corrosion tests on 0·05–0·06% C 18/8/Ti commercial steels

It should be emphasized that the foregoing comments on the superiority of the lower carbon content apply to austenitic chromium nickel steels in the form of welded equipment exposed to onerous corrosive conditions of the 'nitric acid' type. In many chemical plant applications, however, austenitic chromium nickel steels of 18/8/Ti compositions are employed for purposes of cleanliness rather than resistance to 'nitric acid' corrosion and for such applications 18/8/Ti steels containing 0·10–0·13% C are quite satisfactory.

The extensive corrosion testing which has been involved in evaluating the effect of the carbon and titanium contents of the 18/8/Ti steels has been supplemented by the examination of the microstructure of a large number of the corrosion-test specimens. In general these examinations have not yielded much significant information in respect of microstructural features to account for the differences in corrosion resistance, an observation which is not surprising in view of the complexity of the structure of the wrought 18/8/Ti steels. In the particular research work described in this paper, corrosion testing has shown itself to be a useful research 'tool' in that it has revealed certain characteristics of the material not readily detectable by any of the more usual methods of examination.

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The Tempering and Nitriding of Some 3% Chromium Steels

By C. C. Hodgson, A. Met., F.I.M., and H. G. Baron, M.Sc.

SYNOPSIS

The response to nitriding of a series of steels of the En 40 type has been investigated, with particular reference to the effect of previous heat treatment and of carbon and vanadium content.

The steels contained 3% Cr, 0·5% Mo, 0·10–0·36% C, and up to 0·41% V. Small test pieces were first hardened and then tempered for periods of from 4½ min to 100 h at temperatures varying from 500° to 700° C. They were then nitrided for 72 h at 500° C. Hardness readings were taken after tempering and depth/hardness curves were plotted after nitriding.

It was found that the nitrided case-depth and max. case-hardness decrease with decreasing core-hardness (i.e. increasing tempering times and temperatures) except when the core-hardness approaches its minimum value. This decrease in case-hardness and depth is especially serious in the higher-carbon steels. Vanadium helps to counter this effect.

The initial tempering experiments produced results of some intrinsic value and these are presented and discussed separately. The results show the influence of tempering time and temperature on the hardness of this series of steels. 1124

THE STEELS originally used for industrial nitriding were of the well known chromium–aluminium type. It happens that the hardness attained on nitriding such material is almost independent of the tempering-temperature used (at least for steel of medium-carbon content), and consequently of its pre-nitriding hardness. Perhaps because of this the belief arose that the case-hardness after nitriding was governed by the chemical composition of the steel and was independent of variations in the preliminary hardening and tempering operations. It is a view which has been expressed on numerous occasions in the literature. That this view was incorrect was shown by Hodgson and Waring,[1] who examined a number of chromium, molybdenum, and vanadium steels, and established that some aluminium-free nitriding steels, depending mainly on chromium for their nitriding properties, had their response to nitriding influenced considerably by the preliminary tempering treatment.

To explain this behaviour, they suggested that if an alloying element is capable of forming both carbide and nitride, the maximum hardness achieved on nitriding a steel containing it would be governed to some extent by the initial partition of the element between the carbide and ferrite phases. Such partition would be influenced by the tempering-temperature and tempering-time, and by the carbon content of the steel. Detailed examination was restricted to the first of these factors and to a single representative from each of three different types of chromium-nitriding steel.

The present work was undertaken to investigate the influence of tempering-temperature, duration of tempering, and the carbon-content of the steel, on the response to nitriding of steels differing essentially only in carbon content. The steels used were all of the 3% Cr, 0·5% Mo type, and the carbon-contents ranged from 0·10 to 0·36%. The effect of small additions of vanadium was also examined. The tempering experiments which were an essential preliminary to the nitriding experiments furnished results which seemed interesting enough to be included.

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MATERIALS

The chemical compositions of the steels are shown in Table I. Each of the 3% chromium steels was the product of a 1-cwt cast of high-frequency steel. The steels were cast into tapered ingot moulds and the ingots were cogged down from approx. 4 in. sq. to 2½ in. sq. by the steel maker. The bars were then forged down to 1⅛ in. dia. and normalized from 950° C.

A medium-manganese steel (reference *J*) was included in the initial tempering experiments for comparison with the chromium steels. This material was supplied in the form of 1 in. dia. black bar and details of its history are not known.

TEMPERING EXPERIMENTS ON HARDENED SPECIMENS

Thirty-six disc-shaped test pieces were prepared as shown in Fig. 1 from each of the steels. All the specimens were then heated to 900° C in a ' neutral ' salt

Table I

COMPOSITIONS OF STEELS USED, %

Code Letter	C	Si	Mn	S	P	Ni	Cr	Mo	V
H	0·10	0·14	0·53	0·014	0·012	0·06	2·89	0·61	0·02
A	0·20	0·24	0·56	0·014	0·025	0·06	2·82	0·53	0·01
B	0·31	0·30	0·55	0·013	0·005	0·12	2·74	0·52	0·01
C	0·36	0·29	0·65	0·014	0·011	0·10	2·90	0·48	0·02
D	0·25	0·26	0·56	0·015	0·013	0·10	2·96	0·69	0·23
E	0·26	0·28	0·56	0·015	0·006	0·10	3·11	0·61	0·41
J	0·37	0·22	1·33	0·040	0·040	0·34	0·15	0·07	Nil

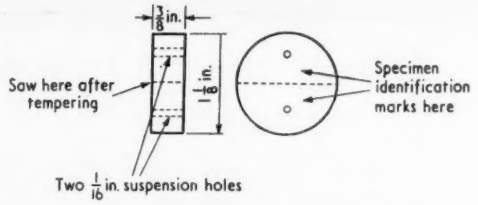


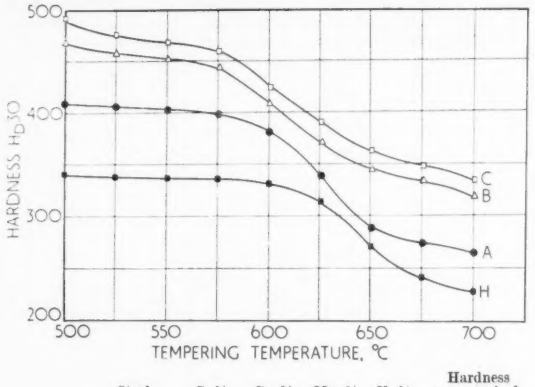
Fig. 1—Dimensions of test piece

bath. The bath was maintained at this temp. for 1 h and the discs were then removed and quenched: the 3% chromium specimens were quenched in oil and the 1·3% manganese specimens were water-quenched.

After cleaning, two test pieces of each steel were hardness-tested. The impressions were made on flats ground on the peripheries of the discs, care being taken to avoid excess heating during grinding. The averages of these hardness readings are shown in Table II.

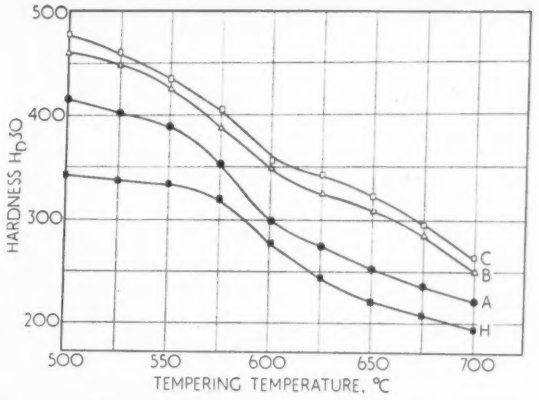
The test pieces were then tempered. Nine temperatures were employed within the range 500–700° C, with increments of 25° C, and four tempering-times were used, 6 min, 1·3 h, 10 h, and 100 h. The times refer to total furnace times. The actual times at the tempering-temperatures were 4·5 min, 1·0 h, 9·7 h, and 99·7 h respectively. The 6-min tempering tests were carried out in a lead bath and the tempering tests of longer duration were carried out in a resistance-wound muffle furnace. All the specimens were cooled in air after tempering.

The discs were then cut in two (Fig. 1) and the sawn faces polished. Diamond-pyramid hardness readings were taken on the polished faces and for each test piece the average of eight or ten readings was found. The relationship between hardness and tempering-temperature has been plotted for the four tempering times in Figs. 2–9. With three independent



Steel	C, %	Cr, %	Mo, %	V, %	Hardness as quenched
A	0·20	2·82	0·53	0·01	440
B	0·31	2·74	0·52	0·01	588
C	0·36	2·90	0·48	0·02	666
H	0·10	2·89	0·61	0·02	370

Fig. 2—Effect of tempering temperature on hardness of Cr–Mo steels with varying carbon contents. Total tempering time 6 min, time at temperature 4½ min.

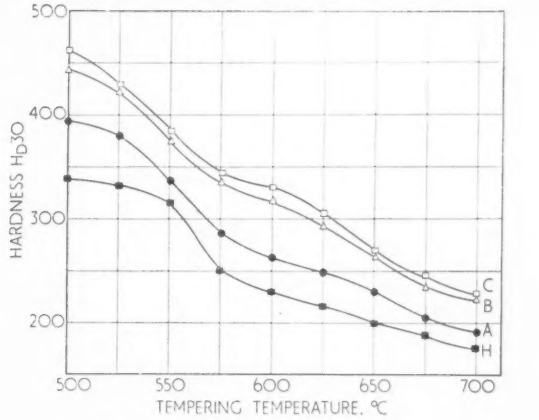


Key as for Fig. 2.

Fig. 3—Effect of tempering temperature on hardness of Cr–Mo steels with varying carbon contents. Total tempering time 1½ h, time at temperature 1 h

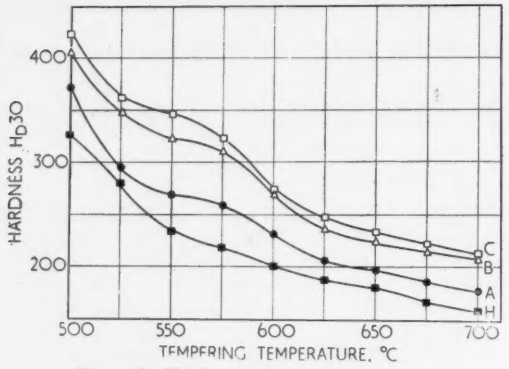
variables the results can be plotted in several different ways which have their separate merits. Two graphs are given in Figs. 10 and 11 which illustrate the effect of tempering time very clearly.

In spite of the complexity of the tempering curves it was found that in each steel the combinations of time and temperature for a given hardness obeyed the well known equation $t = Ae^{(Q/RT)}$, where t is the tempering time and T the temperature, ° K. This equation was perhaps first used in tempering experiments by Bailey and Roberts.[2] Two examples are shown in Figs. 12 and 13, in which $\log_{10}t$ has been plotted against $1/T$, and it will be seen that a good linear relationship is obtained. The activation energy Q varied only slightly with change in hardness, and its value was approximately the same for all the steels. In spite of this it was found that the results fitted the tempering parameter proposed by Hollomon and Jaffe[3] with quite close agreement. Hollomon and



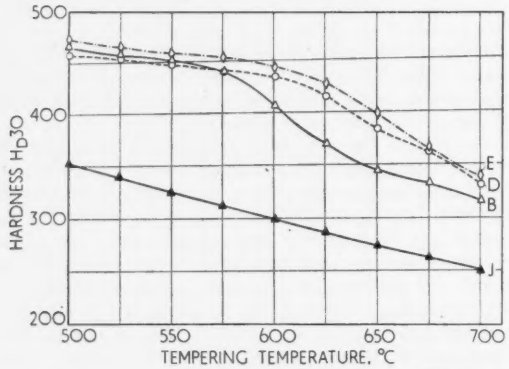
Key as for Fig. 2

Fig. 4—Effect of tempering temperature on hardness of Cr–Mo steels with varying carbon contents. Time at tempering temperature 10 h



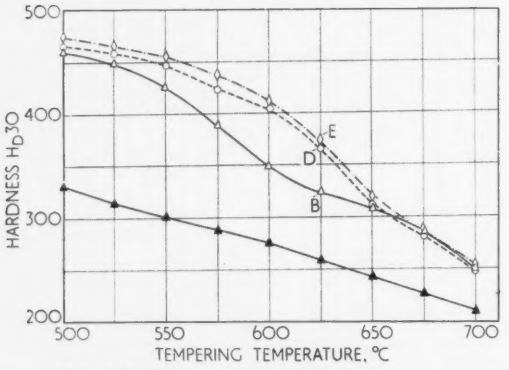
Key as for Fig. 2

Fig. 5—Effect of tempering temperature on hardness of Cr–Mo steels with varying carbon contents. Time at tempering temperature 100 h



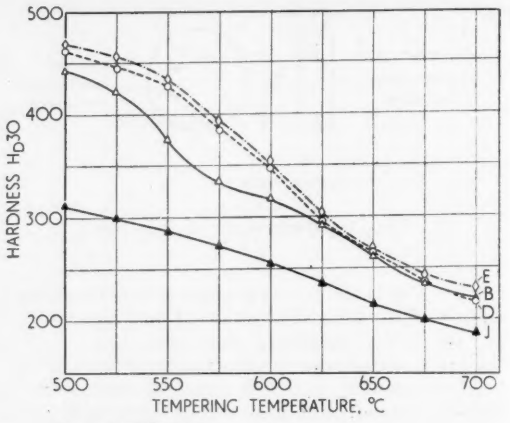
Steel	C, %	Mn, %	Cr, %	Mo, %	V, %	Hardness as quenched
B	0·31		2·74	0·52	0·01	588
D	0·25		2·96	0·69	0·23	513
E	0·26		3·11	0·61	0·41	525
J	0·37	1·33				620

Fig. 6—Effect of tempering temperature on hardness of Cr–Mo–V steels with varying vanadium contents. Total tempering time 6 min, time at temperature 4½ min



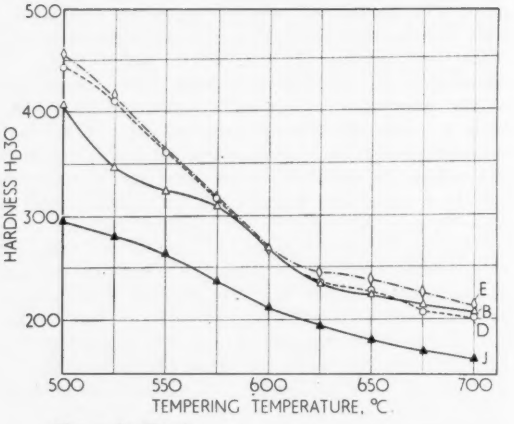
Key as for Fig. 6 ▲ Steel *J*

Fig. 7—Effect of tempering temperature on hardness of Cr–Mo–V steels with varying vanadium contents. Total tempering time 1⅓ h, time at temperature 1 h



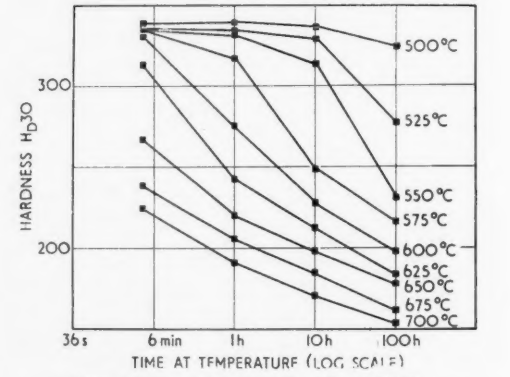
Key as for Fig. 6

Fig. 8—Effect of tempering temperature on hardness o Cr–Mo–V steels with varying vanadium contents. Time at tempering temperature 10 h



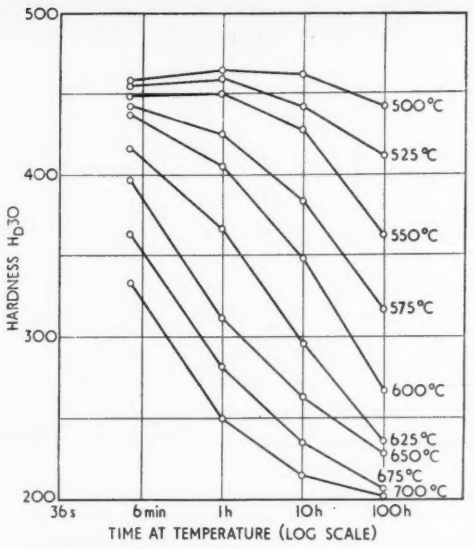
Key as for Fig. 6

Fig. 9—Effect of tempering temperature on hardness of Cr–Mo–V steels with varying vanadium contents. Time at tempering temperature 100 h



Analysis: 0·10% C, 2·89% Cr, 0·61% Mo, 0·02% V
Hardness as quenched: 370

Fig. 10—Steel *H*: effect of tempering time on hardness



Analysis: 0·25% C, 2·96% Cr, 0·69% Mo, 0·23% V
Hardness as quenched: 313

Fig. 11—Steel *D*: effect of tempering time on hardness

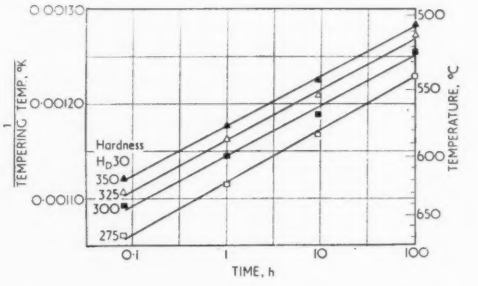


Fig. 12—Steel *A* : relationship between tempering time and temperature for a specific hardness

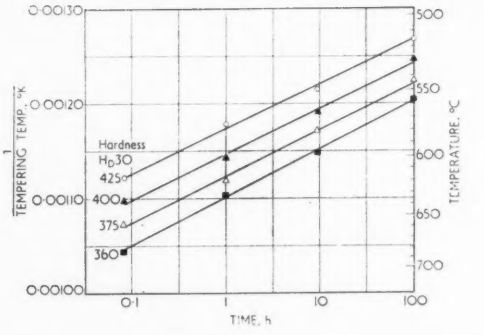
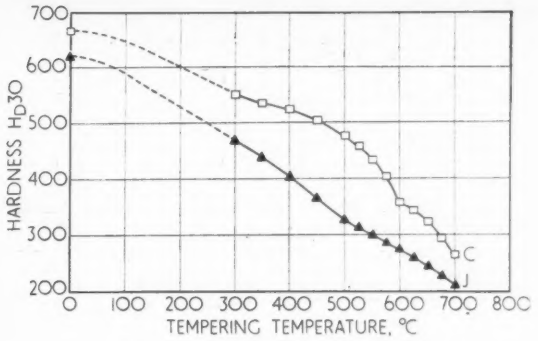


Fig. 13—Steel *D*: Relationship between tempering time and temperature for a specific hardness



Steel	C, %	Mn, %	Cr, %	Mo, %
C	0·36		2·90	0·48
J	0·37	1·33		

Fig. 14—Effect of tempering temperature on hardness. Total tempering time 1½ h, time at temperature 1 h

Jaffe plotted hardness against a parameter which is proportional to Q in the above equation, whilst A was assumed constant. Equally good agreement was obtained with the present results when hardness was plotted as a function of A whilst Q was assumed constant.

The higher carbon Cr–Mo steels show two distinct regions of retarded softening. These regions move to lower temperatures when the tempering time increases, but both can be seen in Figs. 3 and 4. Each retardation is followed by a region of accelerated softening in which the rate of softening with temperature is actually greater than in the 1·3% Mn steel. Finally, at the higher temperatures and when tempering is sufficiently prolonged, the softening rates of all the steels settle down to the same slope.

When a small amount of vanadium is added to the Cr–Mo steels, the part between the two regions of retarded softening is bridged and the hardness in this part is increased by about 50 units (D.P.N.). An increase in vanadium content from 0·23 to 0·41% has little further effect, and it seems probable that the hardening temperature of 900° C was too low to take all the vanadium carbide into solution, especially in the case of the higher vanadium steel.

A few additional tests have been carried out using lower tempering temperatures (Fig. 14). These results confirm that in steels of similar carbon content the

Table II

AS-QUENCHED HARDNESSES FOR 3% Cr–Mo STEELS AND 1·3% Mn STEEL (*J*)

Steel	C, %		Hardness, $H_D/30$
H	0·10		370
A	0·20		440
B	0·31		588
C	0·36		666
D	0·25	0·23 V	513
E	0·26	0·41 V	525
J	0·37	1·33 Mn	620

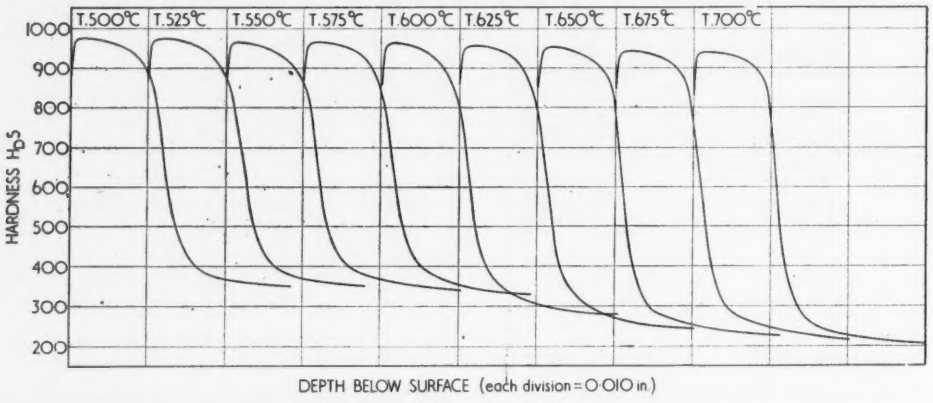


Fig. 15—Depth/hardness curves for steel *H*. Time at tempering temperature 1 h, nitrided 72 h at 500° C

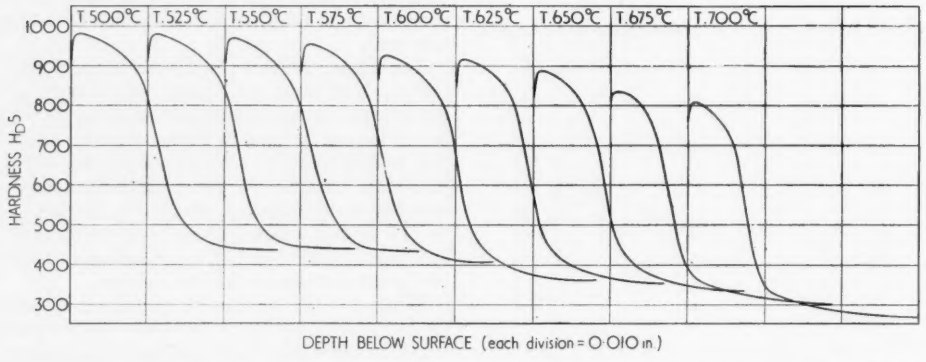


Fig. 16—Depth/hardness curves for steel *C*. Time at tempering temperature 1 h, nitrided 72 h at 500° C

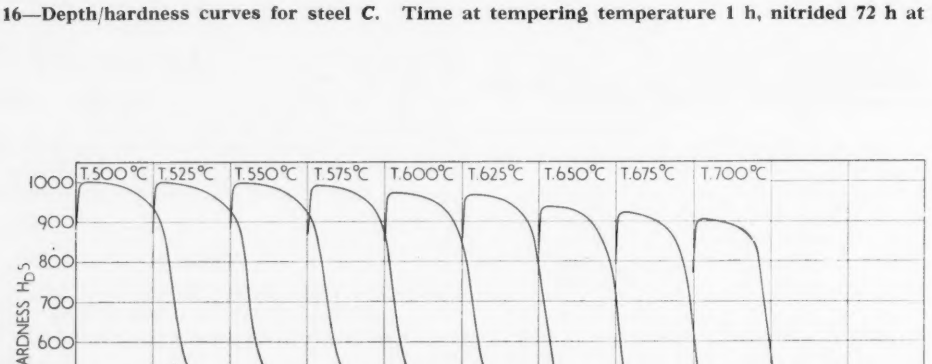


Fig. 17—Depth/hardness curves for steel *D*. Time at tempering temperature 1 h, nitrided 72 h at 500° C

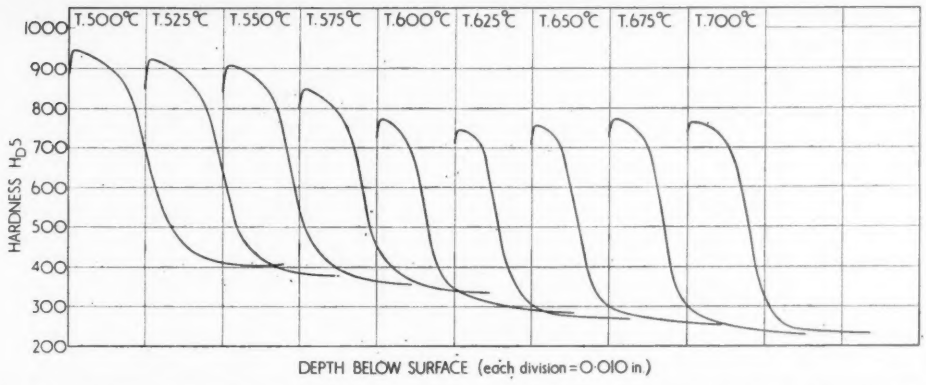


Fig. 18—Depth/hardness curves for steel *C*. Time at tempering temperature 100 h, nitrided 72 h at 500° C

presence of chromium and molybdenum has a considerable effect on the hardness even after tempering at temperatures as low as 300° C. At these low temperatures the hardening effect of the alloying elements is presumably largely due to solid-solution hardening, since it appears from published work that the carbides of chromium and molybdenum have only been detected after prolonged tempering at temperatures above 500° C.

Baluffi, Cohen, and Averbach[4] observed that in chromium steels the spheroidization of Fe_3C is retarded, the particles of Fe_3C remaining small and plate-like and the hardness being retained at higher tempering temperatures than in plain-carbon steels. This phenomenon probably accounts for the first period of retarded softening in Fig. 4.

Each type of carbide can contain considerable proportions of other metallic elements. Kuo[5] has given data on the effect of tempering-time at 560° C on the composition of the carbides in a steel containing 0·51% C and 3% Cr. The concentration of chromium in Fe_3C was found to increase from the matrix value of 3% to a saturation value of 20% in 10 h, after which the Fe_3C was replaced by Cr_7C_3, accompanied by a rapid increase in the concentration of chromium in this carbide. Comparing these results with Fig. 4 it seems probable that the first period of accelerated softening occurs when the chromium content of the cementite is increasing rapidly, and the second retardation of softening coincides with the appearance of Cr_7C_3.

Crafts and Offenhauer[6] have investigated the types of carbides formed in a number of Cr–Mo steels during tempering, using a tempering period of 64 h. Interpolating in their diagrams for a composition of 0·4% C, 3% Cr, and 0·5% Mo (similar to steel *C* in the present experiments), one finds Fe_3C below 550° C, $Fe_3C + Cr_7C_3$ between 550° and 560° C, and Cr_7C_3 above 560° C. With only 0·1% C (steel *H*) the behaviour is similar up to 560° C, but Mo_2C and Cr_7C_3 occur between 560° and 600° C and Cr_7C_3 occurs above 600° C. These results also suggest, when compared with Fig. 5, that the appearance of Cr_7C_3 coincides with the second period of retarded softening. In the low-carbon steel where the molybdenum carbide would be found there is only a suggestion of this second period of retarded softening.

Baluffi, Cohen, and Averbach,[4] using metallographic and other methods, have found a slower rate of replacement of Fe_3C by Cr_7C_3. Long tempering periods at temperatures over 650° C were required for this change to approach completion in a steel containing 0·67% C and 4% Cr. There is also some

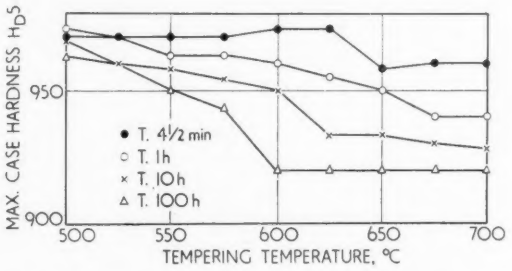
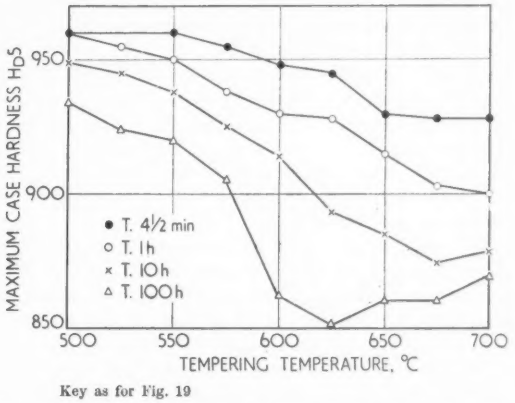


Fig. 19—Steel *H*: influence of tempering before nitriding on hardness of case after nitriding



Key as for Fig. 19

Fig. 20—Steel *A*: influence of tempering before nitriding on hardness of case after nitriding

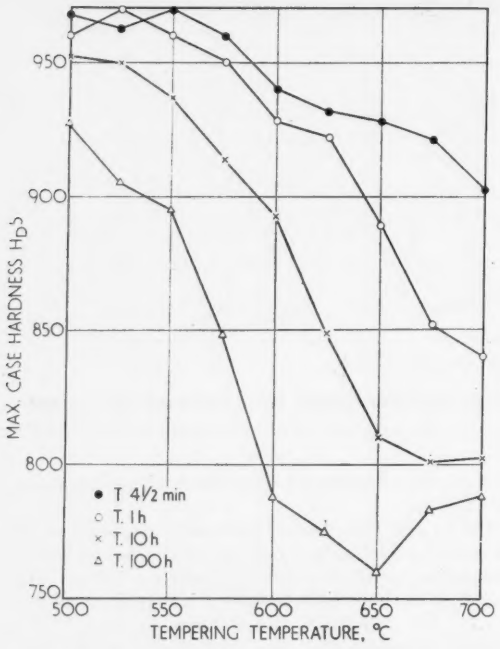


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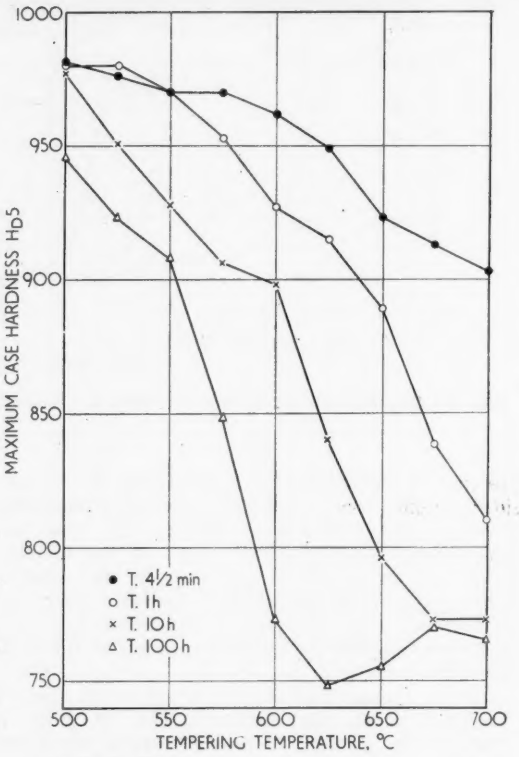
Fig. 21—Steel *B*: influence of tempering before nitriding on hardness of case after nitriding



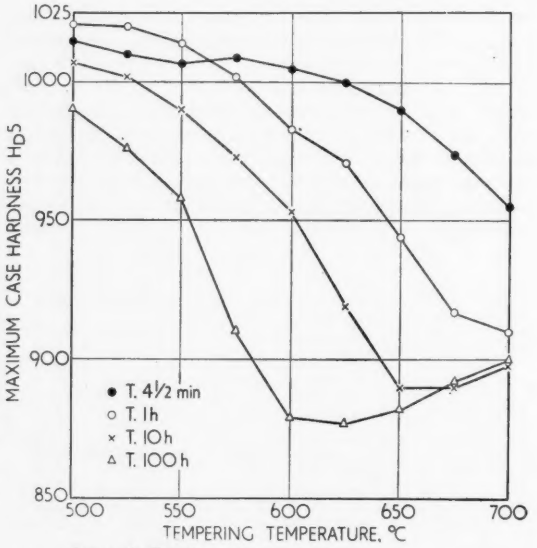
Key as for Fig. 19

Fig. 22—Steel *C*: influence of tempering before nitriding on hardness of case after nitriding

uncertainty about the mechanism of formation of Cr_7C_3.[4, 5]

Both vanadium and molybdenum in sufficient quantities will produce secondary hardening during tempering, with a definite increase in hardness,[7] and it is accepted that this is a result of a precipitation-hardening process. It seems rather doubtful, from the work of Crafts and Lamont,[8] whether the vanadium carbide V_4C_3 would be identified in steels *D* and *E* (Table I), and it has been mentioned that the molybdenum carbide Mo_2C was not found by Crafts and Offenhauer in a steel similar to *C*. Yet both the vanadium and molybdenum undoubtedly help to retard softening in these steels. It may well be, as Crafts and Lamont[9] have suggested, that extremely fine carbide particles are not separated from the matrix by electrolytic extraction, or if separated are not identified by X-ray methods. Alternatively these alloys may retard the growth of Fe_3C (and Cr_7C_3) in the same way that chromium has been observed to retard the growth of Fe_3C.[4]

It is well known that after severe tempering the size and distribution of the carbide particles largely govern the hardness. This point was emphasized when the microstructures of a number of specimens were examined. It was found for instance that the microstructure of the manganese steel *J*, tempered for 1 h at 700° C (D.P.N. 209), was almost indistinguishable from that of a specimen of steel *C*, tempered for 100 h at 700° C and having the same hardness.



Key as for Fig. 19

Fig. 23—Steel *E*: influence of tempering before nitriding on hardness of case after nitriding

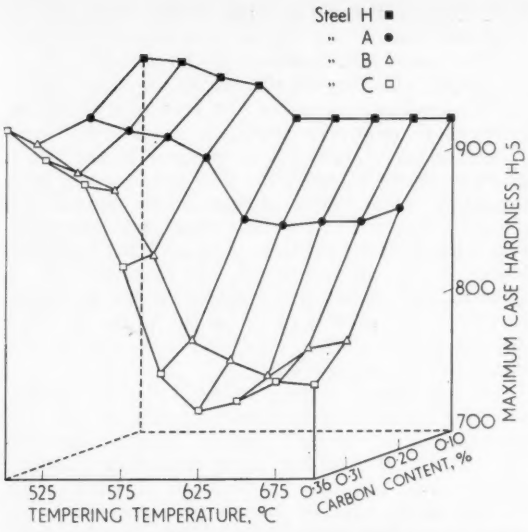


Fig. 24—Relationship between tempering temperature before nitriding, carbon content, and hardness of nitrided case. Tempering time 100 h

NITRIDING OF HARDENED AND TEMPERED SPECIMENS

The sawn faces of the test pieces (Fig. 1) had been ground and polished to 00 grade emery paper and a flat was then ground on the periphery of each specimen, parallel to the polished face, to facilitate taper-grinding after nitriding. Two complete sets of specimens were available and these were nitrided separately in different furnaces. Each set was degreased and then suspended from wires on a steel frame to give free access for gas circulation.

Nitriding was carried out in industrial furnaces using containers of about 70 cu. ft. capacity.[10] The furnaces were soaked at the nitriding temperature of 500° C before receiving the containers, and the total nitriding time, including heating up, was 72 h. The furnace-heating curves and ammonia-dissociation values were similar to those given by Hodgson and Waring,[1] and similar methods have also been used in preparing the nitrided specimens for hardness testing and for measuring the angle of taper after taper-grinding.

Hardness readings were taken on the duplicate sets of nitrided specimens and depth/hardness curves were plotted. Each curve was drawn through the mean of at least two and in some cases up to six sets of hardness readings. Four sets of curves which illustrate most of the features observed are given in Figs. 15–18.

The effects of tempering temperature and time on the maximum hardness of the nitrided case are shown in Figs. 19–23 for steels *H*, *A*, *B*, *C*, and *E*; the combined effects of carbon content and tempering temperature on the maximum case-hardness are illustrated in Fig. 24. The low-carbon steel *H* shows a slight loss of maximum case-hardness when the tempering temperatures and times are increased (Fig. 19). This loss in case-hardness amounts to no more than 30 or 40 D.P.N., and for the longest tempering time no further reduction in hardness occurs when the tempering temperature is increased beyond 600° C. With increasing carbon content the effect of previous tempering conditions becomes more pronounced, and a minimum hardness appears in the 100-h curves at 625° C or 650° C. The case-hardness falls rapidly

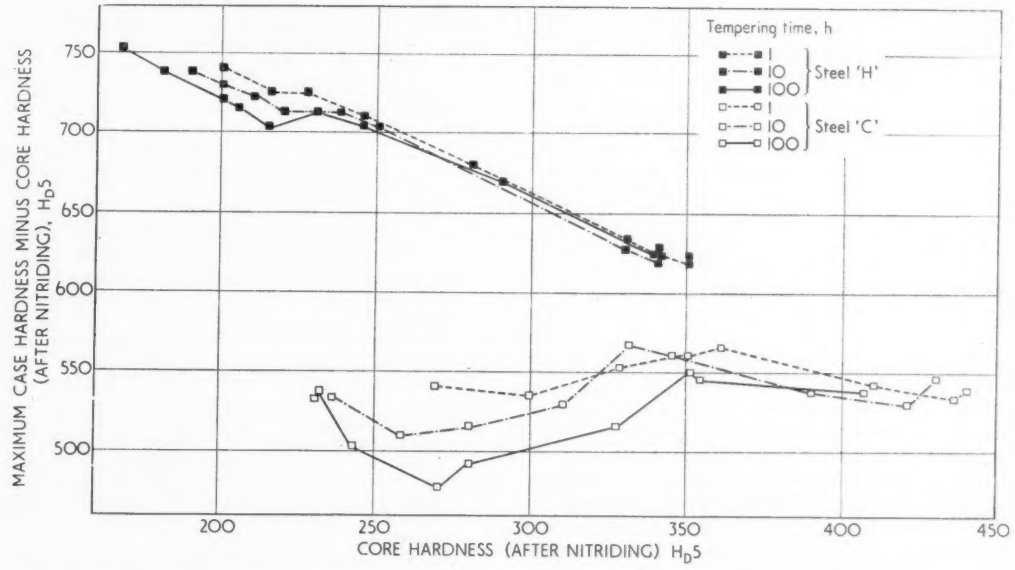


Fig. 25—Increase in hardness resulting from nitriding plotted against core hardness

when tempering temperatures above 600° C and intermediate tempering-times are used.

The increase in hardness obtained by nitriding is plotted against core-hardness for the two steels *H* and *C* in Fig. 25. For periods of tempering usual in industrial practice the results could be represented by straight lines, without introducing serious errors. The low-carbon steel *H* gives the greatest increase in hardness, and the slopes of the curves change from negative to positive as the carbon content is increased.

Direct comparison of the nitriding results obtained from the vanadium-bearing steels *D* and *E* with the vanadium-free steels is not possible because of the differences in carbon-content, but it is quite clear that the inclusion of vanadium increases the case-hardness. The results obtained with steel *D* did not differ greatly from those with steel *E*.

In general, the specimens of a given steel having a similar core-hardness, whether produced by long periods of tempering at a low temperature or short periods at a higher temperature, give similar depth/hardness curves after nitriding. However, there is one marked exception to this rule; it has been pointed out that the case-hardness reaches a minimum value in specimens tempered for 100 h at 625° or 650° C. This suggests that when long tempering periods are used, a slight increase in the concentration of chromium in the ferrite takes place when the temperature is increased above 625° or 650° C.

Kuo and Hultgren[11] have investigated the distribution of alloying elements between ferrite and carbide particles, but their work does not provide any evidence on this point. However, confirmation was obtained from some results by Baluffi, Cohen, and Averbach[4] on the chemical composition of carbides extracted from a 0·67% C, 4% Cr steel, after tempering at temperatures between 538° and 760° C. With increasing tempering temperature (or time) the chromium content of the carbides increased rapidly at first and then more slowly to a limiting value of about 41%. Simultaneously the percentage by weight of carbides in the steel decreased steadily until the Fe_3C was completely replaced by Cr_7C_3. The percentage of chromium in the steel which was contained in the carbide particles can easily be obtained from the data, and hence the percentage of chromium in the ferrite can be found. Because of the different rates of change mentioned earlier, it is found that the amount of chromium in the ferrite decreased rapidly at first, reached a minimum value, and then increased slightly. The minimum value was 0·63% chromium, attained by tempering for 28 h at 649° C, whereas 28 h at 704° C gave 0·78% and 100 h at 760° C gave 0·97%.

Baluffi *et al.* pointed out that the decrease in the proportion of carbides in the steel with increasing tempering is to be expected, if all the carbon is present in the carbides, since Cr_7C_3 (containing 9·0% C) progressively replaces Fe_3C (containing 6·7% C). They also pointed out that these carbon contents remain substantially constant despite variations in the iron/chromium ratio in each carbide, because the atomic weights of iron and chromium are quite similar.

The maximum hardness of the nitrided case has been considered, but the depth of the case is also an important factor. It will be seen from Figs. 15–18 that the case-depth decreased when the case-hardness decreased. This suggests that in the conditions giving a low case-hardness either the rate of diffusion of nitrogen is relatively small or a relatively large concentration of nitrogen is required to produce a given increase in hardness. The latter seems more probable. The higher carbon steels showed the greatest variation in both case-depth and case-hardness with prenitriding treatment, and the combined effect of reduced case-depth and hardness can be quite serious. An interesting feature is that the higher carbon steels showed more acute peaks in their depth/hardness curves than the lower-carbon and vanadium-bearing steels.

The influence of heating to 500° C for 72 h during nitriding on the core-hardness depends on the steel composition and previous thermal history. The low-carbon steel *H* exhibited slight hardening in nearly all instances. The three Cr–Mo steels of higher carbon content were softened if the previous tempering-temperature had been low and the time had been short. Softening became more pronounced as the carbon content increased, but for the same steels there was a tendency for the core-hardness to increase slightly after nitriding if the preliminary tempering had been prolonged and the temperature high.

CONCLUSIONS

(1) In tempering experiments on a series of steels of the En 40 type, using tempering temperatures of 500–700° C and times of 4½ min to 100 h, it was found that:

(i) The higher-carbon Cr–Mo steels show two distinct regions of retarded softening, when the hardness is plotted against tempering temperature

(ii) The first retardation of softening probably occurs whilst the chromium content of the cementite is still small, and the second retardation probably coincides with the appearance of Cr_7C_3

(iii) The ' gap ' between the two regions of retarded softening is bridged if a small proportion of vanadium is added, and a more uniform softening rate is thus obtained

(iv) In each steel the combinations of tempering time and temperature to obtain a given hardness obey an activation energy relationship.

(2) The hardened and tempered samples were then nitrided for 72 h at 500° C and it was found that:

(i) The maximum case-hardness and the case-depth tends to decrease when the previous tempering time or temperature is increased

(ii) The magnitude of this effect depends on the carbon content of the steel. In low-carbon steels the effect is small, but with over 0·3% C it can be quite large

(iii) An exception to 2 (i) occurs when very long tempering times are used. Thus specimens tempered for 100 h at 600–650° C give smaller case-hardnesses and case-depths than specimens tempered at higher temperatures, owing to an increase in the amount of chromium left in solution after tempering at the higher temperatures

(iv) The net increase in hardness caused by nitriding is very approximately a linear function of the core-hardness. The slope of the line changes from negative

to positive when the carbon content of the steel is increased

(v) The results confirm that when the chromium in these steels is segregated into carbide particles the response to nitriding is poor.

Acknowledgments

The authors wish to express their thanks to the Directors of Leyland Motors Ltd., for permission to publish this paper.

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Letter to the Editor

The Reducibility of Iron-Ore Lumps

IN CONTINUATION of the work done on the reducibility of iron-ore lumps,[1] the effect of the shape of a lump on its reducibility was investigated. The hematite ore from the Asswan district in Egypt, which is of an oolitic nature with a rather uniform texture, was used. The total porosity of the specimens was 42%, with a pore diameter of 2·0μ. The chemical analysis of the fairly large lump used for the investigation was: 80·12% Fe_2O_3, 8·01% SiO_2, 2·60% Al_2O_3, 3·99% CaO, 0·86% MgO, 0·16% MnO, 0·01% TiO_2, 2·62% P_2O_5, 0·03% S, and a loss on ignition of 1·46%. The ore resisted thermal spalling.

Lumps of the following geometrical shapes, the smallest dimension being essentially the same, were reduced with hydrogen at 800° C: sphere (diameter = 25 mm); short cylinder (length = diameter = 25 mm); cube (edge = 25 mm); long cylinder (length = twice diameter = 50 mm); long rectangular block (length = twice edge = 50 mm). The experimental set-up and details were similar to those previously described by the writer.[1] The time required for 90% reduction to be accomplished was very nearly the same for the different shapes (65–69 min) and that required for total reduction (99%) was almost twice as much.

Such a result can be easily explained on the basis of the following findings in the study of interfacial movements within a solid ore lump:

(i) The ore is reduced in a topochemical manner.[2-4]

(ii) The advance of the metal/oxide interface, which is the slowest and thus controlling, is at a constant rate.[4]

It follows that the time required for total reduction of a certain lump of ore would be the time necessary for the metal/oxide interface to move through half the smallest dimension of the lump.

This also suggests the use of such a constant rate of advance of the metal/oxide interface (i.e. in millimetres per minute), as an index to the reducibility of the ore under certain experimental conditions. Such an index would have an easily visualized physical meaning.

It is to be noted that there is a certain lag in the first appearance of a metal/oxide interface[4] and this, if not negligible, must be taken in consideration.

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The Straining of Metals by Indentation including Work-Softening Effects

By G. H. Williams, Ph.D., B.Sc., and Hugh O'Neill, D.Sc., M.Met.

SYNOPSIS

The process of elastic and plastic indentation is reviewed, and strain-hardening examined by explorations upon and below indentations in steel, aluminium, and copper. The main study has been on annealed copper using conical tools of three different angles, and indentations of increasing depths made with a ball.

Lubricated cones gave indentations in copper having local Vickers hardness values approaching 124 kg/mm². In the 120° conical cavity the hardest zone was at its centre. Ball indentations at L/D^2 values ranging from 1·96 to 8·85 showed maximum strain-hardening towards the edges of the indents. As L/D^2 was increased to 50 the hardening distribution became irregular but approached a maximum value of 125 kg/mm². These distribution results are contrary to the findings of Krupkowski,[1] both for various cones and for a ball indentation at $L/D^2 = 25$.

Progressive cold-rolling experiments gave a maximum Vickers hardness for copper of 126 kg/mm² followed by a drop to 98 kg/mm². The present work shows that at the higher angles of ball indentation, variations in the results may be complicated by local softening effects. The subject is discussed from the work-softening point of view. 1137

THE DISTRIBUTION OF STRAIN below a loaded indenting tool is very complex, and has recently been studied for limited amounts of ball indentation by Krupkowski.[1] He reported that the maximum strain hardening of the specimen occurs "directly under the top of the ball, and then the hardness gradually decreases." Such a conclusion seemed doubtful and the present paper reviews the indentation process and gives detailed strain-hardening results.

STRAINING BY INDENTATION

The indentation of a large block of metal by any rigid tool involves friction effects and may comprise the following three stages which occur at increasing pressures :

(i) Elastic indentation, up to the yield (flow) stress of the metal
(ii) Plastic indentation with some elastic recovery (as in ordinary hardness tests), proceeding up to immersion of the shoulder or equator of the tool
(iii) Punching, where the shoulder of the tool penetrates deeply and may pierce the metal.

The stress required for stage (iii) was called the pressure of fluidity by Ackermann (1923), and it reaches a steady value when the tool has penetrated to a depth of about five times its 'diameter.' The work of Bishop, Hill and Mott[2] on copper gives the values reported in Table I.

For 'absolute' or 'elastic' hardness the repeated-annealing method of Hanriot may be used, but for metals initially in a metastable condition the procedure of Mahin and Foss[3] might be employed, the indentation being machined or ground until it fits its loader indenter. Alternatively the punch-yield-strength test of Hanemann[4] could be considered, but his yield values are all higher than his Brinell

Manuscript received 28th March, 1955.
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Table I

RESULTS FOR 3·75-in. DIA. BARS OF COPPER[2]

	Annealed, kg/mm²	Work-hardened, kg/mm²
Yield stress in compression	4·7	27·6
60° cone hardness, 'lubricated'	...	63
" " " unlubricated	...	116
" " " unlubricated punching pressure	132	187

Table II

ELASTIC OR 'ABSOLUTE' HARDNESS (kg/mm²)

Procedure and Reference	Material	Standard Brinell No.	Mean Pressure 'Absolute' Hardness	Ratio
10-mm ball[6] and repeated annealing	Al bronze	80	38·5	2·1
" [7]	71·2% Cu–Zn	65	19·4	3·3
	60·9% Cu–Zn	80	32·8	2·4
" [8]	Mild steel	95	68·4	1·4
" [9]	Ingot iron	79	54·3	1·5
	Malleable	111	73·5	1·5
	0·12% C steel	116	63·0*	1·8
	0·95% C "	195	122·0	1·6
" [10]	Copper	42	13·6	3·1
120° cone and repeated annealing[10]	"	42	13·6	3·1
Machined impression[3]	Copper	42	17·1	2·5
"	Bronze	56	31·0	1·8
"	Iron	83	47·0	1·8

* Approximate

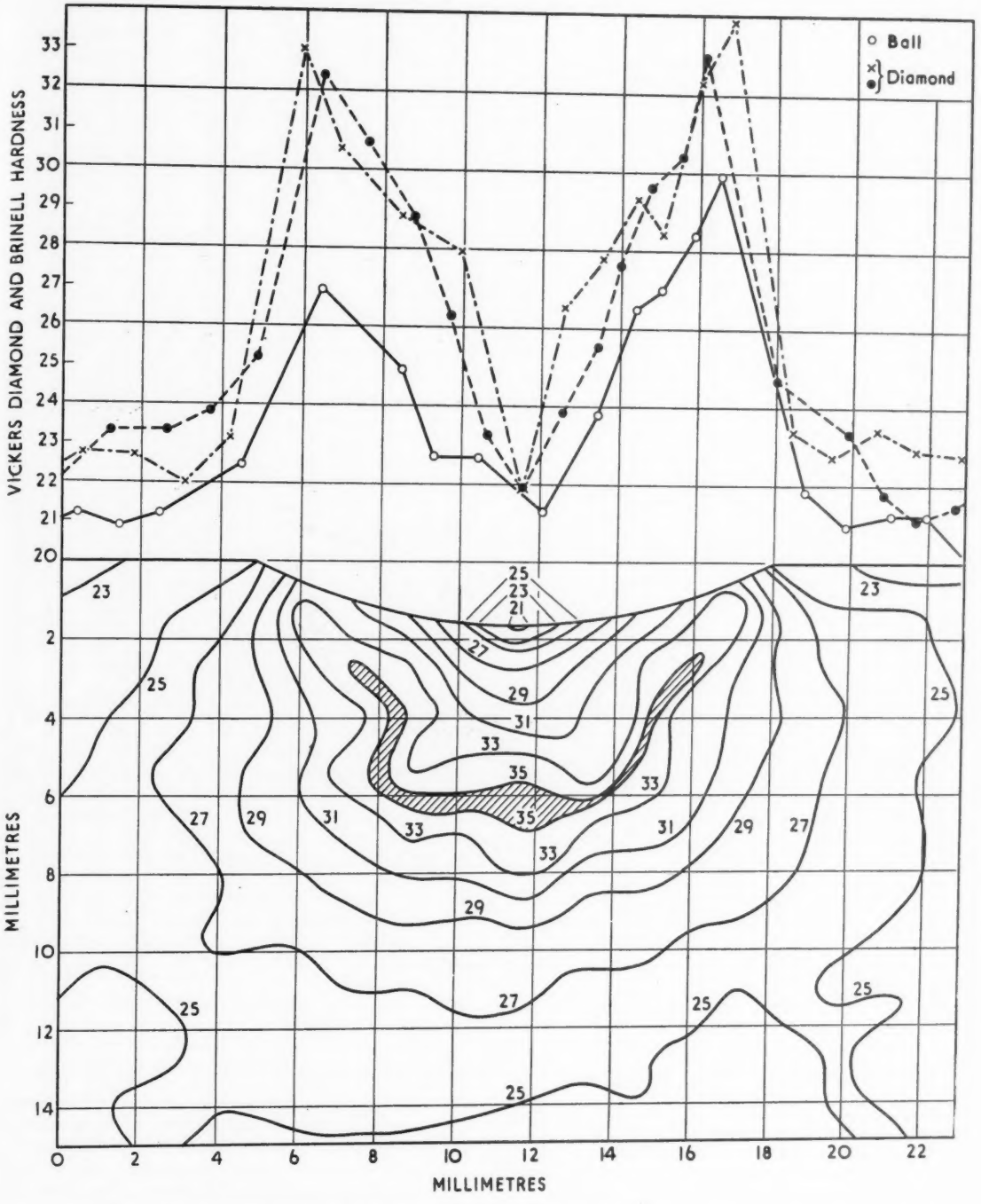

Fig. 1—Distribution of strain hardening in unlubricated indent of aluminium. Surveys of section, and of surface in 2 directions at right angles, were made by diamond pyramid test. A surface survey was also made by the 1-mm steel ball

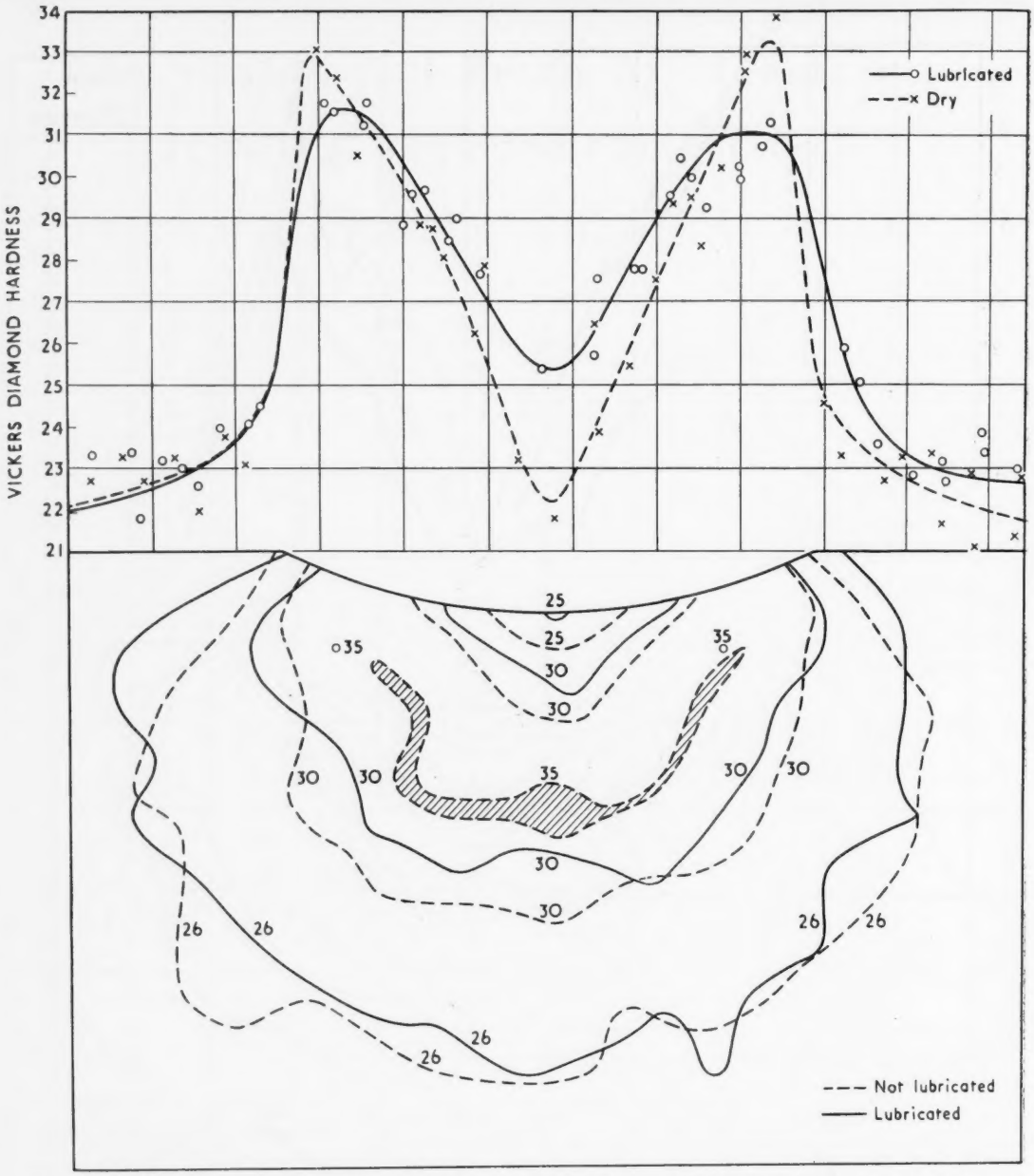

Fig. 2—Comparison of strain-hardening distribution in lubricated and unlubricated indent in aluminium

values, and therefore have not been included in the results which we have collected in Table II.

The elastic hardness will exceed the yield stress of a compression test because of the supporting effect of the mass of metal surrounding the indentation. The value of 13·6 for copper (Table II), divided by the yield stress of 4·7 (Table I) gives a ratio of 2·9. The theoretical work of Hencky and Ischlinski shows a ratio of 2·8, whilst a value of 2·23 on cold-rolled copper was found by Bishop, Hill and Mott.[2]

Researches by photo-elastic and other methods have revealed the strain distribution for stage (i), indentation by a ball. Theoretically, if the circle of contact has a diameter d, the maximum shear stress is greatest at a depth of about $d/4$ below the centre of the contact circle, and plastic deformation begins when Pressure $(P_m) = 1{\cdot}1 \times$ (yield stress).

HARDNESS DISTRIBUTION IN PERMANENT INDENTATIONS

Krupkowski[10] has explored with a 1-mm ball test the hardness on the contact surfaces of large indenta-

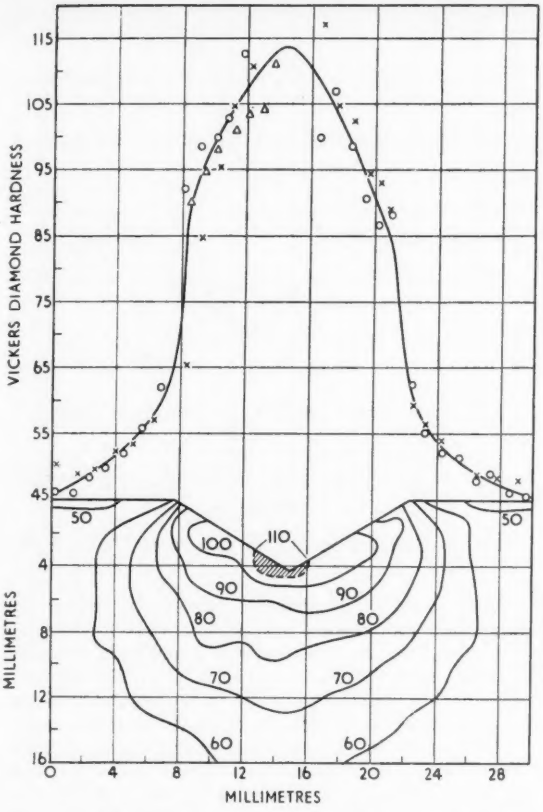

Fig. 3—Indentation in copper with 120° cone at 7825 kg

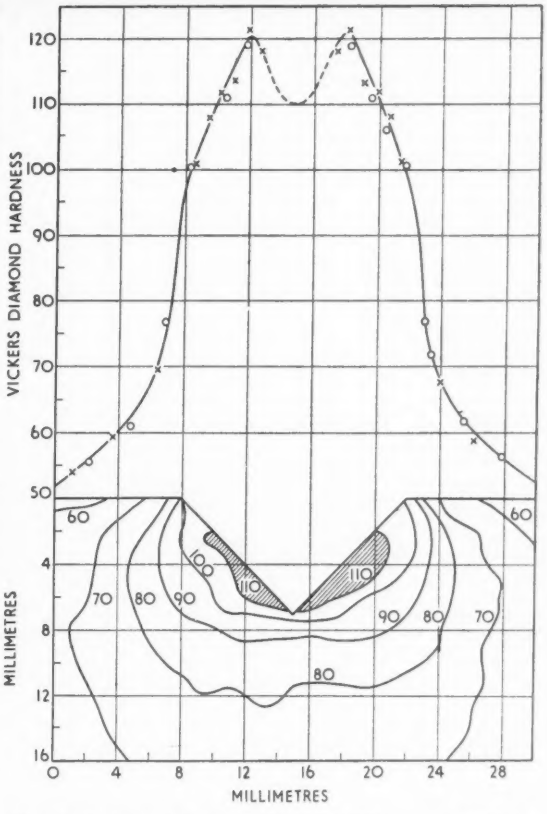

Fig. 4—Indentation in copper with 90° cone at 8280 kg

tions, and reported that the metal in contact with large conical indenters delevoped a uniform value over the surface—apart from a slight decrease as the apex of the indent is approached. As the angle of the testing cone was decreased higher surface-hardness values were reported. For ball indentations at $L/D^2 = 25$, on copper, surface hardness was a maximum at the centre and decreased towards the rim.

These results seemed doubtful and one of the authors, with the assistance of Mr. J. Dearden, M.SC., made some preliminary experiments early in 1947. Values obtained for ball and pyramid cavities are given in Table III, the large pyramid of Widia being loaned by the British Cast Iron Research Association. The effects for the lubricated indentations are contrary to those obtained by Krupkowski, but the surface-hardness gradient decreased with increase of L/D^2, and presumably might vanish and even reverse.

Further tests were started early in 1948, and meanwhile Tabor[12] also showed that the maximum Vickers hardness occurred inside the rim of ball impressions. Exploration of a 10-mm ball indentation ($L/D^2 = 50$)

Table III

INDENTATIONS IN 60/40 BRASS (94 BRINELL, n = 2·35)

Indenter	L/D^2	Indent Dia. or or Diagonal, mm	Max. Pyramid Hardness (5 kg load)				
			Surface of Indent			Sectioned Specimen	
			Inside Rim	At Centre	Difference, %	On Centre-line	Depth Below Surface, mm
27-mm ball	10·0	10·0	153	120	−27	...	...
	16·7	12·5	158	125	−26	145	2·5
	30·0	16·0	167	143	−17	162	1·8
136° pyramid	3000-kg load	8·0	139	180	+30	170	0

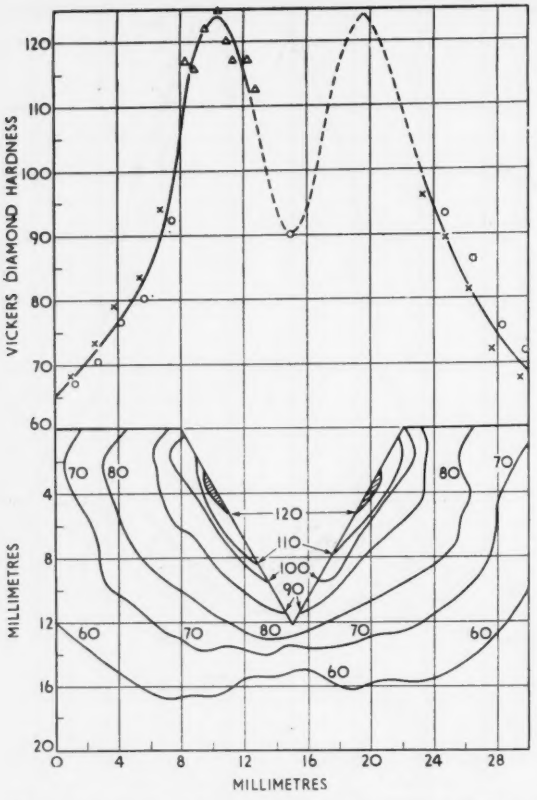

Fig. 5—Indentation in copper with 60° cone at 8640 kg

in mild steel (C 0·25%, Mn 0·53%) gave a surface hardness of 230 D.P.N. inside the rim, with a fall to 160 at the centre. At a depth of about 2 mm below the surface on the centre-line, the hardness reached 200 D.P.N.

EFFECT OF LUBRICATION ON BALL INDENTATIONS IN ALUMINIUM

With the object of studying the effect of lubrication on hardness distribution, tests were made on aluminium (99·5%). A 28·55-mm ball degreased in benzene and ethyl alcohol was used with a load of 2500 kg ($L/D^2 = 3·08$) for the dry conditions. An attempt was made to obtain the same size of indentation with repeated lubrication and this only required a load of 2350 kg, contrary to the work of Hankins who did not observe much effect of lubrication for the ball test. Explorations with a diamond pyramid and a load of 5 kg and also with a 1-mm ball at 5 kg load are given in Fig. 1 for the dry condition. Results for the pyramid tests are shown in Fig. 2, the drop in hardness at the centre of the impression being considerable, with the value at this position only one unit higher than that of the unstrained aluminium. The pyramid test gave higher values than the Brinell and is preferable for these explorations as it yields hardness values which are strictly comparable with each other.

The numerical difference between the maximum and minimum on the surface-hardness curve for the lubricated indentation is less than that for the dry specimen, but their position was found to be the same in both cases for this value of L/D^2. The hardness distribution is again contrary to the results of Krupkowski whatever conditions of lubrication he used.

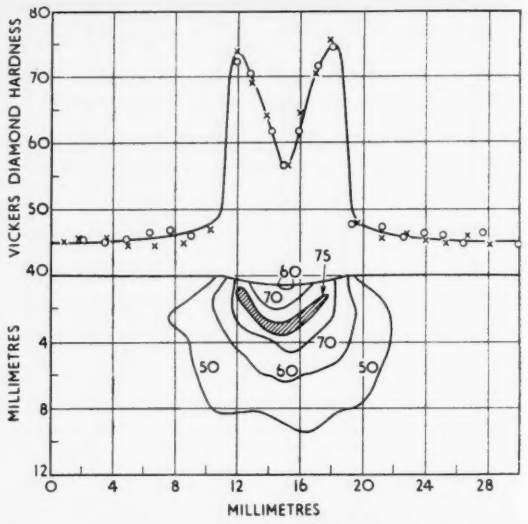

Fig. 6—Ball indentation in copper at $L/D^2 = 1·96$

Differences in the sectional surveys are probably associated with stagnant cap effects under the ball. A cup-shaped zone of maximum hardness 35 is located at about 5 mm below the centre of the dry indentation, and is replaced by a ring of $H_D = 35$ in the lubricated specimen.

EXPLORATION OF CONE INDENTATIONS

Special conical indenters, having the usual included angles of 60°, 90°, and 120°, were specially made from high-speed steel with highly polished surfaces and diamond pyramid hardness values of 800–900. Sharper cones might have had theoretical interest but are seldom used and show greater differences for the lubricated and unlubricated conditions.[2] Indentations were made on annealed copper cylinders of 99·93% purity on a Buckton lever machine at loads required to give indents of 14 mm in all cases. Liquid paraffin B.P. was used as a lubricant by repeated applications after unseating the indenter. Two hardness surveys were carried out on the surface of indented specimens across two diameters at right angles, using the Vickers machine at 10 kg load.

The hardness distribution for the 120° cone is similar to that for the 136° pyramid (Table III) as the maximum hardness value is at the apex of the indentation and decreases steadily towards the rim. As the cone angle is decreased the hardness maximum is increased slightly and displaced towards the edge of the cavity (Figs. 3, 4, and 5).

The isosklers on these sectioned specimens run in from the horizontal surface so as to be concave towards the indentation. Savitskyi, Vandyshev, and

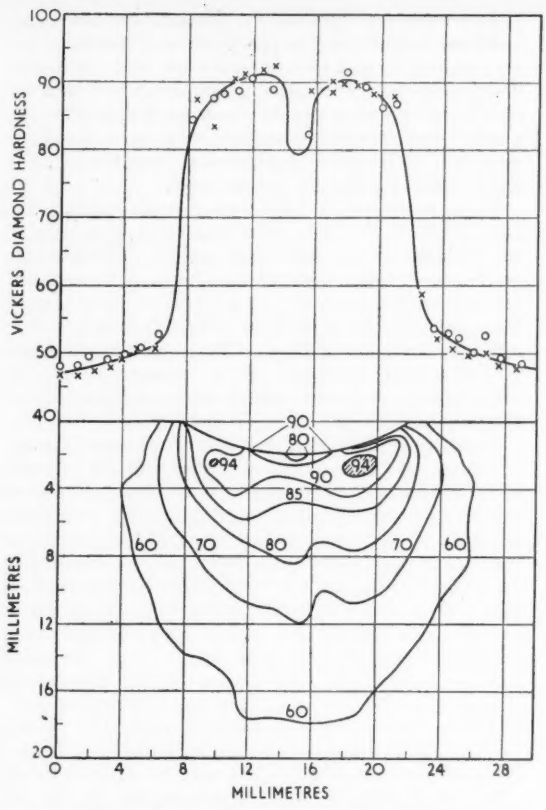

Fig. 7—Ball indentation in copper at $L/D^2 = 8{\cdot}85$

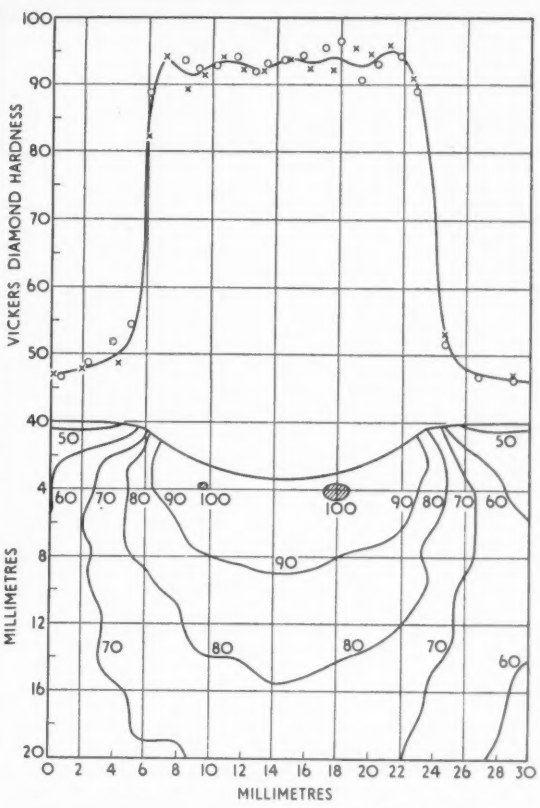

Fig. 8—Ball indentation in copper at $L/D^2 = 15$

Yakutovich[13] explored indentations made in a block of alloy steel of about 260 Brinell with a 90° cone, and in their case the hardness contour lines as they run from the surface of the block are seen to be convex to the cavity and then become concave as they approach the centre-line. This difference is probably associated with the ' sinking-in ' of copper at the rim of the indents.

BALL TESTS ON COPPER

Large wrought specimens from the same cast of annealed copper were used for this work, indentations with a 28·55-mm polished steel ball being made at L/D^2 ratios from 1·96 to 50 using repeated lubrication with paraffin. The blocks were large enough to meet B.S. requirements for indentation testing. The results of surface and sectional explorations with the diamond pyramid are shown in Figs. 6–10, the former being completed within two days of indentation and the latter within six.

At the lower loading ratios the surface hardness (Vickers) is a minimum at the centre of the indentation and rises to a maximum towards the rim. At $L/D^2 = 15$, it is practically uniform, and by $L/D^2 = 25$ the maximum approached 110 with variations in the graph. These continue at higher loading ratios, and hardnesses of 125 were reached. Various contours have been collected in Fig. 10.

Some cold-rolling experiments were made on pieces of the annealed copper, and Vickers tests completed within 5 min of each reduction. Figure 11A shows the work-hardening curves thus obtained. A maximum of 126 at 92% reduction was obtained, followed by a drop in Vickers hardness to 98 at 98% reduction. The work of Cook and Richards[14] has shown that a double fibre texture is produced by this degree of rolling and also that softening by self-annealing can take place under certain conditions.

Curve B in Fig. 11 gives the general mean pressure hardness values against the corresponding angles of indentation produced by the large ball. Similarly the values for the three cones have been plotted. The maximum surface hardness graphs, when extrapolated to the limiting angle of indentation (180°), rise to about the same maximum as on the cold-rolling graph. This is roughly the punching pressure for (different) copper recorded in Table I, and is about twice the ultimate hardness obtained in the present ball tests. Cylinders of our copper compressed in this laboratory on lubricated platens by Dr. Polakowski gave a Vickers hardness maximum of 133 kg/mm².

DISCUSSION OF THE RESULTS

Sectional hardness surveys of ball indents in aluminium and copper showed that for low values of L/D^2 the region of maximum hardness is a cup-like zone at

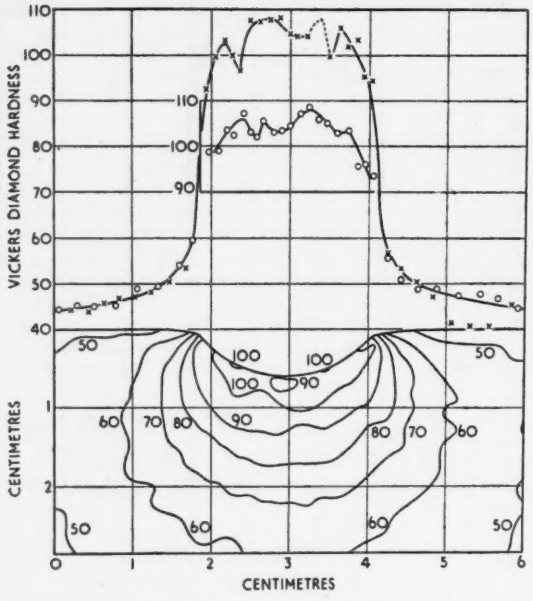

Fig. 9—Ball indentation in copper at $L/D^2 = 25$, including repeat survey of surface of cavity

about $d/3$ below the surface of the indents. However, for high values of L/D^2 on copper the resulting heavy deformation caused the 'cup' to be replaced by two maxima on the section, representing an internal ring of superhardened metal. At the higher loading ratios the ring moved towards the surface of the indent, which then reached a fairly uniform hardness. Eventually the surface hardness varied between 125 and 110 as immersion of the ball to its equator ($d = D$) was approached.

The change on the surface is probably due to a work-softening affect induced by heavy strain. Concerning the absolute distribution of the hardening or softening, it must be remembered that the exploratory tests themselves add a further amount of local strain, and that residual elastic internal stresses in the specimen can affect the size of the indents produced during this local testing. Other work has shown that with internal elastic tension the exploratory hardness value may be reduced by about 5%.[15]

The strain-hardened zones are always within the depth of specimen required by B.S. 240 : (1926), but would exceed a thickness suggested by Weingraber and Barghoorn.[16]

Reference may be made here to the suggestion of Franke[5] for the determination of elastic or absolute hardness by deducting from the usual Brinell number the additional hardness at the centre of the impression measured by the Herbert pendulum or the Shore scleroscope. Such a method could only be very approximate because, as we have shown, the surface hardness at the centre of the indentation undergoes complex variations. In general, however, it does appear that, if a ball indent were made so that $d = D$, the highest Vickers value on its surface would represent the strain-hardening limit of the metal.

An alternative method for determining the elastic hardness might use the specimen and indenter as a thermocouple, and plot temperature rise against indenting stress. The initial graph due to friction and elastic compression should show an inflexion at the plastic yield stress. Sensitive apparatus would be necessary as Herbert[17] registered a mean rise of 32° C when making a Brinell test on steel.

The present results differ from Krupkowski's in that the surface hardness is usually never a maximum at the middle of a ball indentation. Furthermore, on traversing the edge of the impression and proceeding outwards along the original surface, there is an extremely sharp drop in hardness which is very different from his findings. We might mention that Krupkowski's specimens were smaller than B.S. requirements and were reinforced by an annular steel ring.

Savitskyi *et al.*[13] used the results of sectional surveys of conical indentations in steel to determine the amount of cold work by compression at various positions in the section. They equated their values with those obtained by indenting the centre of the top surface of cylindrical specimens compressed to various extents. This procedure is of doubtful value as the hardness may vary considerably from point to point on a compression specimen as the work of Polakowski[18] shows. In the early stages of compression, no strain-hardening can be detected at the centre of the compressed faces where these investigators made hardness tests.

There is reason to believe that the actual numerical value of diamond hardness obtained for any particular degree of cold work may be affected by the method of working and the orientation of the distortion effects. Variations in creep rate during the process of indentation can occur, for Hargreaves[19] in his studies of work-softening showed that annealed copper had a flow index S of 0·0026 which reached 0·007 after drawing, whilst ageing for 1 h at 100° C increased S to 0·0138 in the worked metal. For a fixed time of loading the heavily strained material will flow more and therefore give a lower indentation hardness value. Hargreaves notes that during reorganization and recrystallization there are irregulari-

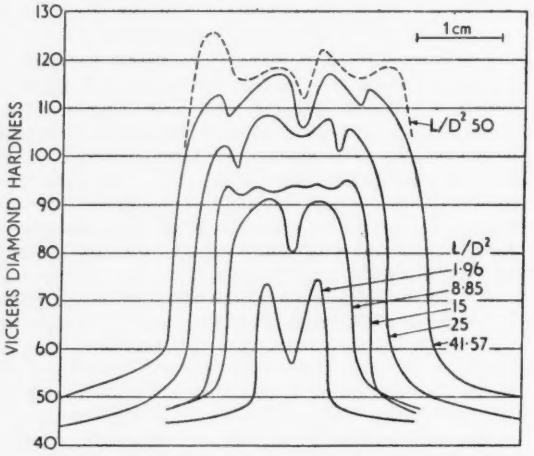

Fig. 10—Surface hardness of ball indentations

ties in the indentation properties. Similarly the irregularities observed at 40–50% reduction during the cold-rolling of copper—which have sometimes suggested a temporary work-softening—are probably due to mixed orientation effects. The observed[20] rise of the Meyer n value from 2·0 to 2·03 with heavy cold-rolling may also be consistent with a work-softening effect, but it has been stated[2] that "n should be slightly greater than 2 even for a material which does not show work-hardening." One might wonder whether Meyer's law will hold at the highest loading ratios for a metal showing these variations, and on plotting our results we find that it does up to $d = D$ (approx.).

Since this work was concluded explorations of strain-hardening below dynamic ball indentations of small angle in copper and steel have been published.[21] It is worth noticing a work-softening and inflexion of the hardness gradient just within the strain-hardened zone. This might be explained by work-softening as observed by Polakowski[22] and ascribed by him to a Bauschinger effect. Polakowski expressed his suggestion in the following terms: "Work softening occurs whenever the method of cold working tends to produce (in the metal) an internal stress system different from that left after the preceding treatment, irrespective of whether the latter was of mechanical or thermal nature."

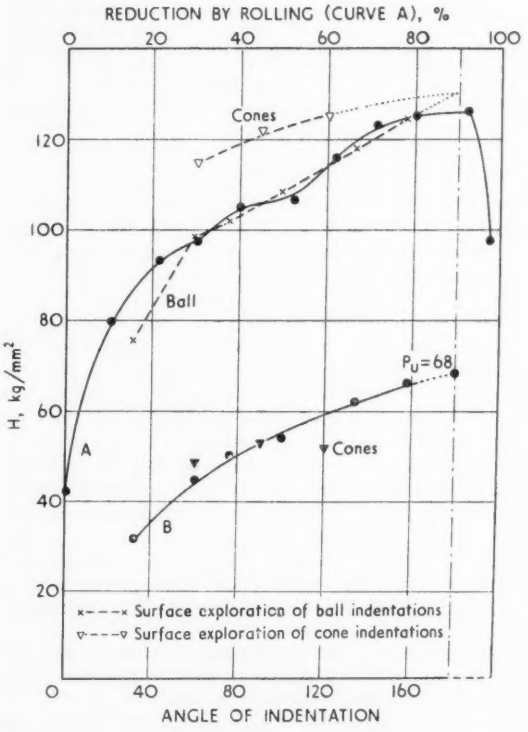

Fig. 11—Maximum surface hardness values on large ball and cone indents. Curve B gives mean pressure hardness values for these indents. Curve A represents Vickers hardness of cold-rolling experiments

Another point of interest is the appreciable decrease of diamond pyramid hardness number of copper and certain steels when tested with increasing loads, as Fisher[23] emphasized. We consider that variable load-support arising from 'concavity' or 'convexity' effects on the edges of the indentations is involved. These depend upon the strain-hardening and displacement of the metal and consequently upon the distribution of internal stresses. We are sympathetic to Fisher's suggestion that certain observed decreases of Vickers number with rise of testing load are due to work-softening, and friction effects may possibly be involved. Fisher also suggested that the work-softening effect was greatest in steels as their carbide-particle size decreased. This is consistent with observations of Polakowski, as he was able to explain a curious result reported by one of us[24] that a sorbitic tyre steel only gives a Vickers hardness rise of 7 kg/mm² when cold-rolled as much as 40% reduction.

Acknowledgment

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The Solubility of Graphite in Manganese, Cobalt, and Nickel

By E. T. Turkdogan, R. A. Hancock, and S. I. Herlitz

SYNOPSIS

The solubility of graphite in pure manganese, cobalt, and nickel has been determined at various temperatures up to 1700° C. The solidus temperatures of binary mixtures of carbon with manganese, cobalt, and nickel are found to be 1308°, 1319°, and 1314° C; the latter two are the eutectic temperatures, whilst the former is thought to be a peritectic temperature. The plots of log (atom fraction of carbon) against the reciprocal of the absolute temperature give good straight lines. There are accurate data available on the solubility of graphite in copper and iron melts; these, together with the present results, indicate that the solubility of graphite in the elements of the First Long Period of the Periodic Table decreases with increase in the atomic number of the elements. The solubility is also related to the second ionization potential of the metals under consideration. 1241

CARBON is one of the elements which has a limited solubility in metals, whether the latter are solid or liquid. Several workers investigated the solubility of graphite in pure molten iron[1–4] and in molten iron alloyed with various elements of the Second Short Period of the Periodic Table.[2–8] The literature surveys show that no work has been carried out on the thermodynamics of carbon dissolved in transitional elements other than iron. Attempts were made by Sykes[9] and Kihlgren and Eash[10] to construct cobalt–carbon and nickel–carbon phase equilibrium diagrams from data of some early work.[11–18] The information on the solubility of carbon in manganese is meagre and the experimental results are conflicting.[3, 19, 20] It was thought desirable therefore to investigate the solubility of graphite in molten manganese, cobalt, and nickel.

EXPERIMENTS

Pure electrolytic manganese, cobalt, and nickel were used. About 5 g of the metal powder, or small beads in the case of nickel, were packed in a small graphite crucible, 2 in. long, 0·4 in. bore, and 0·6 in. outside diameter, which was made from a graphite rod of high purity. The crucible was then placed in the hot zone of a vertical furnace fitted with a 1-in. recrystallized-alumina tube; depending on the temperature of the experiments, a platinum-wound or a molybdenum-wound furnace was used. The temperature measurements were taken by means of a Pt/Pt–13% Rh thermocouple; the thermocouple was kept in a thin-walled recrystallized-alumina sheath which was located in close contact with the base of the graphite crucible. The preliminary measurements indicated that the melt was within the uniform-temperature zone of the furnace and, moreover, temperature readings at the base of the crucible were practically the same as those taken by a second thermocouple immersed temporarily in the melt. An automatic ON–OFF controller maintained the temperature of the furnace well within $\pm 2°$ C.

To avoid oxidation of the metal and crucible, the experiments were carried out in oxygen-free dry nitrogen. At the end of the experiment the crucible was rapidly pulled out of the reaction tube with a molybdenum wire and water-quenched. In the case of melts with manganese, only the bottom half of the crucible was immersed, making sure that the melt was not in direct contact with water; this arrangement prevented the decomposition of manganese–carbon specimens. Traces of free graphite adhering to the surface of the metal were removed by grinding and rubbing with a fine emery cloth. The specimens were then reduced to a fine granular state in a percussion mortar.

The carbon contents of the samples were determined by the combustion method, agreement between duplicates being within $\pm 0·01\%$.

A few preliminary experiments, at 1350° and 1500° C, indicated that the metal was saturated with graphite within 1 h of the start of the test. This is in line with some recent work on the solubility of carbon in iron and iron–phosphorus melts.[4] In all

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Dr. Turkdogan is Head of the Physical Chemistry Section of the Chemistry Department, Mr. Hancock is in this Section, and Mr. Herlitz was with the Association as a Vacation Student from the Royal Institute of Technology, Stockholm, Sweden.

Table I

SOLUBILITY OF GRAPHITE IN MOLTEN MANGANESE, COBALT, AND NICKEL

No.	Temp., °C	Carbon	
		wt-%	N_C
In Manganese			
A 1	1350	7·54	0·272
2	1400	7·55	0·272
3	1450	7·82	0·280
4	1500	8·01	0·285
5	1600	8·26	0·292
6	1650	8·52	0·299
7	1700	8·67	0·303
In Cobalt			
B 1	1350	2·73	0·121
2	1400	2·90	0·128
3	1450	3·02	0·133
4	1500	3·23	0·141
5	1550	3·28	0·143
In Nickel			
C 1	1350	2·09	0·095
2	1450	2·45	0·109
3	1500	2·68	0·119
4	1550	2·83	0·125

the subsequent runs, 2 h were given for the establishment of equilibrium between the pure metal and the graphite crucible.

Metallographic examination of some of the nickel and cobalt samples indicated that graphite flakes and nodules, which were precipitated during quenching, were dispersed evenly across a section of the sample. It is worth noting that whilst carbon-saturated manganese, iron, and cobalt samples were very brittle (brittleness becoming less as the atomic number of the base element increases) the carbon-saturated nickel samples were found to be very hard to saw and drill and very tough to break. Although a careful way of quenching prevented the instantaneous decomposition of manganese–carbon melts, these samples decomposed slowly on exposure to atmosphere. In view of this, the samples from manganese melts were analysed for carbon immediately after quenching and crushing of the samples and in no circumstances were the samples allowed to be wetted with water.

RESULTS

The experimental results are given in Table I, where the composition in the last column is in terms of atom fractions. In Fig. 1 $\log N_C$ (where N_C is the atom fraction of carbon at saturation) is plotted against the reciprocal of the absolute temperature; the points lie on good straight lines within the temperature range studied. For comparison, the line representing the temperature dependence of the graphite solubility in pure iron has been reproduced from a previous publication of Turkdogan and Leake.[4]

To facilitate the calculation of carbon saturation

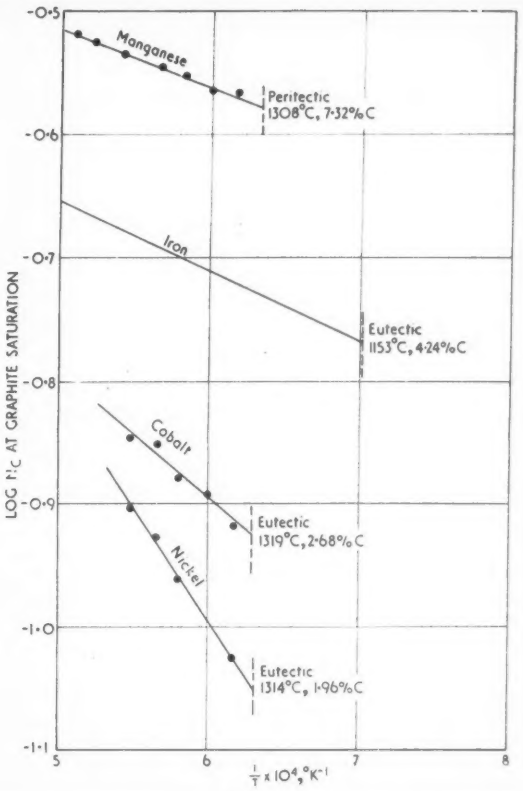

Fig. 1—Solubility of graphite in some transitional elements

at any temperature the following expressions, derived from the lines in Fig. 1, may be used:

Temperature dependence of graphite solubility in

(*a*) *Manganese*

$$\log N_C = -\frac{460}{T} - 0{\cdot}285 \quad \ldots\ldots(1)$$

(*b*) *Iron*

$$\log N_C = -\frac{560}{T} - 0{\cdot}375 \quad \ldots\ldots(2)$$

(*c*) *Cobalt*

$$\log N_C = -\frac{1050}{T} - 0{\cdot}265 \quad \ldots\ldots(3)$$

(*d*) *Nickel*

$$\log N_C = -\frac{1820}{T} + 0{\cdot}097 \quad \ldots\ldots(4)$$

The solidus temperatures of carbon-saturated manganese, cobalt, and nickel samples were determined by the conventional heating and cooling curves. For these thermal analyses, the thermocouple, protected by a thin-walled graphite sheath, was inserted into the melt which was in a 1-in. dia. graphite crucible. The temperature gap between the arrests on the heating and cooling curves was about 5° C and the values of solidus temperatures recorded in Fig. 1 are the averages of those obtained by heating and cooling.

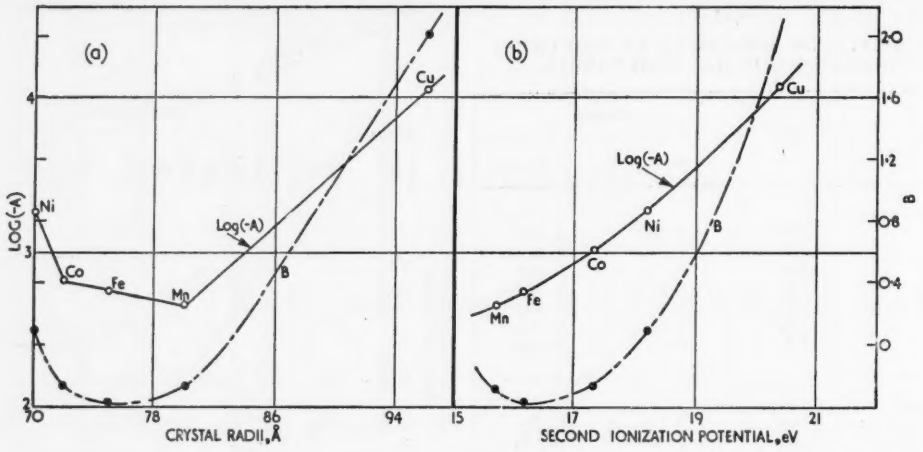

Fig. 2—Log ($-A$) and B related to (*a*) crystal radii and (*b*) second ionization potential of metals in the First Long Period

Whilst the melts of nickel, cobalt, and iron form eutectics with carbon at 1314° C (1·96% C), 1319° C (2·68% C), and 1153° C (4·24% C), respectively, the solidus temperature 1308° C (7·32% C) in manganese–carbon melts is well above the melting point of pure manganese,[21] i.e. 1244° C. In the latter case, therefore, the invariant point may correspond to a peritectic reaction; at this point, the reaction between manganese–carbon melt and β-graphite may form a manganese carbide. There is also the possibility that the higher temperature of the solidus may be due to a melting point maximum of the manganese–carbon solid solution. It was, however, stated by Kuo and Persson[22] that no carbide of manganese is stable above 1100° C. It appears that further work on the thermodynamic properties of manganese carbides is desirable.

DISCUSSION

It is clear from the relationship in Fig. 1 that at temperatures below 1700° C the solubility of carbon in manganese, iron, cobalt, and nickel decreases, in the order written, as the atomic number of the metal increases; beyond the transitional elements, the solubility of graphite in other metals of the First Long Period becomes extremely small. In fact, measurements by Bever and Floe[23] have shown that even at 1800° C the solubility of graphite in pure copper does not exceed 0·006 wt-%. From their results the values of log N_C are plotted against the reciprocal of the absolute temperature and the best straight line drawn through the experimental points may be represented by the following expression:

Temperature dependence of graphite solubility in copper

$$\log N_C = -\frac{11{,}500}{T} + 2{\cdot}014 \qquad \ldots\ldots(5)$$

In equations (1) to (5) the temperature coefficients and the constants will be designated by A and B, respectively. Although the terms A and B are not separately related to the heat and entropy of solution of carbon in the metal, they are nevertheless the characteristics of the temperature-dependence of carbon solubility in various molten metals.

In Fig. 2*a*, log ($-A$) and B are plotted against the empirical crystal radii given by Pauling.[24] Carbon is known to occupy the interstitial sites in these metals and it is reasonable to expect that the atomic size of the solvent should have some influence on the size of the interstitial sites, and hence on the solubility of carbon. The relationship in Fig. 2*a* is rather complex and indicates that the variations in the values of (A) and (B) from one element to another are not controlled solely by the atomic size of the solvent metal. It should, however, be mentioned that although no information, as yet, is available on the atomic diameters of metals in the liquid state, the empirical crystal radii used in Fig. 2*a* should bear some relation to the true atomic radii of the molten metals.

Another factor likely to influence the solubility of carbon in metals is the nature and the magnitude of the ionization of the solvent and solute atoms. It is known that a carbon atom dissolved in austenite loses electrons and becomes positively charged.[25–30] It may be possible that a similar ionic structure (or partial ionic structure) will exist in the liquid state. With this in view the values of log ($-A$) and B were plotted against ionization potentials[31–32] of the elements under consideration. When the first ionization potentials were used, the points did not lie on a smooth curve. For the plots in Fig. 2*b* the second ionization potentials are used. The values of log ($-A$) and B increase continuously (with the exception of the value of B for manganese) with the second ionization potential of the solvent metal in such a way that, within the range 1100–1700° C, the solubility of carbon decreases as the ionization potential of the solvent element increases. Since the elements under consideration belong to the same Period of the Periodic Table, their various physical and chemical properties vary in a systematic manner with their atomic number, and correlation of various properties

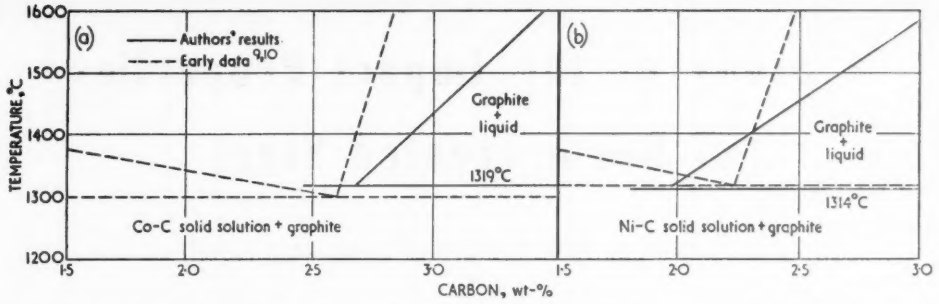

Fig. 3—Comparison of present work with early data on part of the phase diagram in (*a*) Co–C and (*b*) Ni–C binary systems

against one another may produce smooth curves for which no immediate theoretical reason can be assorted. The relationship in Fig. 2*b* may therefore not have a simple theoretical explanation.

The variation of carbon solubility in the transitional elements is also in line with the extent of interaction between metal and carbon atoms measured by the heats and free energies of formation of various carbides. From the data compiled by Richardson[33] it is to be seen that ΔG^0 at 0° K per mean g-atom, for the formation of M_XC compounds, decreases as the atomic number of the element M decreases from Ni (28) to Ti (22), and it is in this order that the solubility of carbon increases at a given temperature.

The findings of the present investigation differ widely from those of early workers[9, 10] as shown in Figs. 3*a* and 3*b*; there is, however, good agreement on the eutectic temperatures of cobalt–carbon and nickel–carbon binary systems.

Chipman *et al.*[3] measured the solubility of carbon in ferromanganese at 1290°, 1490°, and 1690° C. They showed that manganese increases continuously the solubility of graphite in melts containing up to 80% manganese and down to about 10% iron. Extrapolation of their curves to zero iron content gives carbon solubilities comparable with those determined by the authors.

Acknowledgment

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A Study of the Impact Properties of Boron-treated Steel

By Samuel J. Rosenberg and John D. Grimsley

SYNOPSIS

Twenty-one steels from the same open-hearth heat and made to a base composition of 0·43% carbon and 1·60% manganese, were studied to evaluate the effect of boron on the Charpy V-notch impact properties over the temperature range of −140° to + 140° F. The boron which was added in various amounts by means of seven different intensifiers, improved the impact properties of the steels tempered at 400° F but generally appeared to be detrimental to the impact properties of the steels tempered at 800–1200° F and air-cooled. The extent of this detrimental effect depended upon the type of addition agent, the amount of boron, and the tempering temperature. Temper brittleness of the untreated steel (induced by furnace-cooling from 1200° F) was increased when boron was added by five of the intensifiers. Considering the overall effect of boron, it appears that boron may be added in amounts up to 0·0025% in the form of simple intensifiers without significantly affecting the impact properties. 1145

ALTHOUGH BORON-TREATED STEELS have been in use for more than a decade and some of their properties have been well established, no systematic study of the effect of boron on temper brittleness has been reported. This is surprising in view of the exhaustive literature on temper brittleness and the detrimental effect of this phenomenon on the engineering applications of heat-treated steels.

Two comprehensive reviews on temper brittleness are available. Hollomon[1] covered the literature before 1944 and Woodfine[2] made a critical survey of published work from 1944 until 1953. Although no references to boron-treated steels were to be expected in Hollomon's review, it is significant that no such references appear in Woodfine's paper.

Imhoff and Poynter[3] noted that 80B30 steel was susceptible to temper brittleness and Powers and Carlson[4] that susceptibility to reversible temper embrittlement was increased by boron. Udy and Rosenthal[5] reported that boron additions apparently had no adverse effect on notched-bar toughness down to − 80° F, and that no direct effect of boron content on susceptibility to temper brittleness was evident.

Possibly the most comprehensive study of the effect of boron on the properties of steels was made by Digges and Reinhart.[6] The phenomenon of temper brittleness was not reported upon, but they saw that boron appeared to improve the impact properties of steels treated to high hardness levels. At lower hardness levels, boron appeared to be slightly detrimental to the impact properties. Similar observations were noted by Wilcock.[7] Sheehan and Glassenberg[8] also reported that boron-treated steel had superior impact properties at high hardness levels (about Rc 55).

MATERIALS, TREATMENTS, AND TEST METHODS

A series of boron-treated steels, made from one commercial open-hearth heat, was available for this investigation. The preparation and properties of these steels (1 base steel and 20 boron-treated steels) have been described by Digges and Reinhart[6]; the chemical compositions and austenite grain sizes are given in Table I. The ingots were numbered consecutively as they were poured, the highest number being the last ingot of the heat used for test purposes. No boron was added to the fifteenth ingot (C15) in order to have the base steel from near the middle of the pour. Boron was added by means of seven different addition agents, the compositions of which are given in Table II. Since all the ingots were poured from the same heat (the boron additions were made in the mould), all the test steels were essentially of the same base composition and any variations in impact properties or in susceptibility to temper brittleness could be ascribed to the boron or to the other elements present in the addition agents.

As furnished to the laboratory, all the steels were in the form of 1¼-in. round bars, and then they were normalized in nine ½-in. lengths at 1650° F. Total furnace time was 1½ h, followed by air-cooling. Impact specimens, located as shown in Fig. 1, were shaped to about 0·415 in. square and cut to proper lengths before hardening and tempering, after which they were ground to size and notched. The specimens were hardened by quenching from 1550° F (½ h at temperature) in oil and then tempered (1 h at temperature) at 400° (only 9 steels were tempered at this temperature), 800°, 900°, 1000°, 1100°, or 1200° F, followed by air-cooling. For the development of temper brittleness, additional specimens tempered at 1200° F were furnace-cooled at a rate of 30° F/h.

All tests were made in a Charpy impact testing machine of 224-ft-lb capacity, using a standard specimen (0·394 in. square by 2·165 in.) with a V-notch (0·079 in. deep, 0·010 in. root radius). The testing machine conformed to A.S.T.M. specifications apart from the radius of the striking edge; this was 0·047 in. instead of 0·315 in.

Impact tests were conducted over a range of

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Table I

CHEMICAL COMPOSITIONS OF THE TEST STEELS, wt-%

Steel	C	Mn	P	S	Si	Ni	Cr	Mo	V	Ti	Zr	Cu	Al (total)	Nitrogen		Boron			Intensifier	A.S.T.M. Grain Size No. at 1550° F
														Acid soluble*	Total†	Added	Recovered			
																	Acid soluble‡	Total		
C 6	0·44	1·58	0·017	0·019	0·40	0·01	0·05	0·01	Nil	0·019	0·004	0·02	0·040	0·003	0·003	0·0011	0·0009	0·0009	Addition agent A	7, few 5's
C 7	0·43	1·59	0·018	0·016	0·45	0·01	0·05	0·01	,,	0·027	0·005	0·02	0·050	0·003	0·004	0·0015	0·0016	0·0016	,, ,,	8
C 8	0·43	1·58	0·019	0·019	0·50	0·01	0·05	0·01	,,	0·053	0·010	0·02	0·058	0·003	0·003	0·0022	0·0031	0·0031	,, ,,	8
C 9	0·43	1·58	0·019	0·019	0·32	0·01	0·05	0·01	,,	0·020	0·004	0·02	0·030	0·002	0·004	0·0006	0·0003	0·0003	Addition agent B	7–8
C10	0·42	1·60	0·019	0·018	0·32	0·01	0·05	0·01	,,	0·030	0·006	0·02	0·047	0·002	0·003	0·0013	0·0006	0·0006	,, ,,	8
C11	0·42	1·59	0·019	0·018	0·32	0·01	0·05	0·01	,,	0·052	0·007	0·02	0·053	0·002	0·003	0·0019	0·0011	0·0011	,, ,,	8
C12	0·43	1·56	0·020	0·019	0·31	0·01	0·05	0·01	,,	0·001	0·002	0·02	0·049	0·004	0·005	0·0020	0·0009	0·0022	Addition agent C	7, few 5's
C13	0·43	1·58	0·020	0·017	0·31	0·01	0·05	0·01	,,	0·001	0·002	0·02	0·045	0·004	0·004	0·0030	0·0017	0·0028	,, ,,	5–6
C14	0·43	1·59	0·018	0·019	0·32	0·01	0·05	0·01	,,	0·001	0·002	0·02	0·042	0·002	0·005	0·0040	0·0026	0·0036	,, ,,	4–6
C15	0·43	1·58	0·020	0·019	0·31	0·01	0·05	0·01	,,	0·001	0·002	0·02	0·025	0·004	0·004	Nil	0·0000	0·0001	None	8
C16	0·42	1·59	0·019	0·021	0·33	0·01	0·04	0·01	0·004	0·005	0·004	0·02	0·047	0·002	0·003	0·0013	0·0012	0·0012	Addition agent D	7, few 4's
C17	0·43	1·62	0·019	0·019	0·35	0·01	0·04	0·01	0·004	0·013	0·004	0·02	0·054	0·002	0·004	0·0027	0·0025	0·0025	,, ,,	7–8
C18	0·43	1·64	0·020	0·019	0·37	0·01	0·04	0·01	0·004	0·030	0·005	0·02	0·069	0·002	0·003	0·0040	0·0038	0·0038	,, ,,	8
C19	0·43	1·58	0·019	0·018	0·31	0·01	0·04	0·01	0·015	0·004	0·003	0·02	0·040	0·002	0·005	0·0004	0·0001	0·0001	Addition agent E	7–8
C20	0·43	1·58	0·019	0·020	0·31	0·01	0·04	0·01	0·027	0·015	0·002	0·02	0·045	0·002	0·004	0·0008	0·0003	0·0004	,, ,,	8
C21	0·42	1·57	0·019	0·020	0·32	0·01	0·04	0·01	0·070	0·020	0·003	0·02	0·054	0·002	0·005	0·0011	0·0006	0·0006	,, ,,	8
C22	0·43	1·57	0·019	0·021	0·35	0·01	0·05	0·01	0·004	0·005	0·003	0·02	0·031	0·003	0·005	0·0005	0·0004	0·0007	Addition agent F	7–8
C23	0·43	1·57	0·021	0·020	0·35	0·01	0·04	0·01	0·020	0·012	0·005	0·02	0·046	0·004	0·005	0·0010	0·0008	0·0008	,, ,,	8
C24	0·43	1·57	0·019	0·019	0·42	0·01	0·05	0·01	0·025	0·015	0·010	0·02	0·045	0·003	0·005	0·0016	0·0011	0·0011	,, ,,	8
C25	0·43	1·57	0·019	0·021	0·34	0·01	0·05	0·01	0·004	0·001	0·002	0·02	0·048	0·004	0·005	0·0019	0·0007	0·0016	Addition agent G	7–8, few 5's
C26	0·43	1·57	0·021	0·020	0·36	0·01	0·04	0·01	0·004	Nil	0·002	0·02	0·046	0·004	0·004	0·0036	0·0015	0·0026	,, ,,	8

* Soluble in 1 : 1 or stronger sulphuric acid
† Soluble in 1 : 1 or stronger sulphuric acid containing approx. 10 vol-% hydrogen peroxide (perhydrol)
‡ 1 : 1 phosphoric acid

temperatures (from $-140°$ to $+140°$ F) in order to define the transition range. With the exception of those specimens broken at room temperature, all were held in liquid baths at the required temperatures for at least 15 min before testing. The average time required to transfer a specimen from the liquid bath to the anvil of the machine and to release the hammer was about 3 s; during this interval no significant change in temperature occurred in the specimen.

All broken impact specimens were examined visually to evaluate the percentage of brittle fracture. As an aid to this examination, and to reduce to a minimum the human error in estimation, a set of drawings was prepared for use as comparison standards. It is believed that the estimations of the areas of brittle fracture are good to $\pm 5\%$.

Two hardness measurements were made on each of the finished impact specimens; the averaged values of the many determinations made for each steel in each condition of heat treatment are given in Table III. The hardness of the temper-embrittled steels were slightly lower (about 2 points Rc) than the non-embrittled steels tempered at 1200° F; this was caused by the longer sojourn at the high tempering temperatures resulting from slow cooling.

RESULTS AND DISCUSSION

The transition temperature as determined from notched-bar impact tests is not a limiting temperature above which a steel may be used with assurance that brittle failure will not occur, nor is it the limiting temperature below which a steel dare not be used for fear of brittle failure. The transition temperature of any given steel is dependent upon the geometry of the specimen, the depth and sharpness of the notch, the distance between supports, the speed and weight of the striking hammer, the criterion used for selecting

Table II

CHEMICAL COMPOSITIONS OF THE ADDITION AGENTS*

Addition Agent	Composition, wt-%							
	Mn	Si	Ti	Zr	Al	Ca	B	V
A		39·54	9·84	7·49	6·60	9·68	0·44	
B			23·52	5·76	12·22		0·65	
C		2·63					11·56	
D	24·00	20·26	17·76		14·05		1·33	
E		1·22	13·38		10·38		0·38	26·88
F		38·22	9·00	9·11	6·69		0·53	9·49
G		40·21					2·50	

* These agents were ferro-alloys

the transition temperature, etc. Many investigators of the effect of test temperature on the impact properties of steels have used the temperature corresponding to an arbitrarily selected energy level as the transition temperature, even though at the selected value the fracture may be entirely brittle (granular).

As there is no standard method of selecting either the transition temperature or the transition temperature range, the data obtained in this study were evaluated upon the basis of six arbitrary criteria, selected as follows:

(i) Maximum temperature corresponding to 100% brittle fracture
(ii) Temperature corresponding to 50% brittle fracture
(iii) Minimum temperature corresponding to 0% brittle fracture
(iv) Temperature corresponding to 20 ft-lb breaking energy
(v) Temperature corresponding to 40 ft-lb breaking energy
(vi) Temperature corresponding to the average breaking energy between 100% and 0% brittle fracture; this is indicated in the figures as $E = \frac{1}{2}(E_{100\%\,B} + E_{0\%B})$.

For each of these criteria, additional data are of interest, as follows:

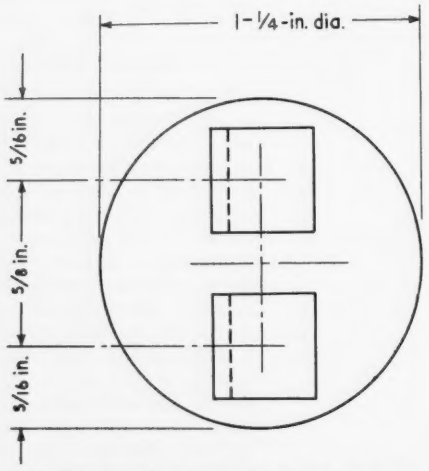

Fig. 1—Sketch showing location of impact specimens with reference to cross section of original bar

Table III

HARDNESS (ROCKWELL C SCALE) OF THE TEST STEELS

Steel No.	Tempering Temperature, ° F						
	400° A.C.*	800° A.C.	900° A.C.	1000° A.C.	1100° A.C.	1200° A.C.	1200° F.C.†
C6	...	40·5	34·5	30·5	26	20·5	18
C7	54	40·5	35	31	27	21	19
C8	...	42	36	32	28	22·5	20·5
C9	...	40	33·5	30	26	20	18
C10	53	39·5	34·5	31	27	20	18·5
C11	54	40·5	35·5	31·5	27	21	19
C12	...	40·5	34	30	25·5	20·5	18
C13	53·5	39·5	34·5	30	25·5	20	17·5
C14	...	40	33	30	26	19·5	18
C15	54	40	35	30·5	27	20·5	18·5
C16	53·5	40	35·5	31	25	20·5	18·5
C17	...	41	35	31	27	22	19
C18	...	40	34	31	27	21	19
C19	...	41	34	31	27	20	18·5
C20	53·5	41	36	32·5	28	22	20
C21	...	41	35	32·5	29	24	22
C22	...	40	34	30	26	20·5	18·5
C23	...	41	35	31·5	28	21·5	20
C24	54	41	36·5	32	28·5	21	20
C25	54	39·5	34	31	26·5	19·5	18
C26	...	40	33·5	30·5	26	20·5	18·5

* A.C. = air-cooled
† F.C. = furnace-cooled

For items (i), (ii), and (iii): the energy absorbed.
For items (iv) and (v): the percentage of brittle fracture.
For item (vi): the energy absorbed and the percentage of brittle fracture.

Effect of Tempering Temperature on Impact Properties

It is not practical to present, either in graphical or tabular form, all the test data obtained. Impact-energy/test-temperature curves were plotted for all the steels and a typical set of curves is shown for the base steel in Fig. 2. Examination of the curves for all the steels showed certain features of interest.

Although not all of the steels were tempered at 400° F, enough were tested in this condition to confirm the conclusion of other investigators that, when treated to high hardness levels, boron has a beneficial effect upon the notched-bar toughness over the test

Table IV

IMPACT VALUES OF THE STEELS AS TEMPERED AT 400° F

Steel No.	Boron, %	Addition Agent	Impact-Energy Absorbed, ft-lb at	
			+ 70° F	− 40° F
C 7	0·0016	A	18·0	13·3
C10	0·0006	B	17·6	15·3
C11	0·0011	B	17·0	14·2
C13	0·0028	C	15·6	13·6
C15	...	None	12·6	6·3
C16	0·0012	D	15·2	13·5
C20	0·0004	E	18·0	15·5
C24	0·0011	F	17·0	13·5
C25	0·0016	G	15·8	12·0

temperature range studied. At least one steel having boron added by each of the seven different intensifiers was tempered at 400° F. Table IV gives the impact value of these steels at test temperatures of + 70° and — 40° F.

As may have been expected, tempering at successively higher temperatures (800–1200° F) effected a continuous increase in notched-bar impact values at room temperature. However this increase in impact toughness was not always maintained at lower test temperatures. In all the boron-treated steels except C13, the curves representing the 900° F temper crossed either or both curves representing the 1000° and 1100° F tempers. This indicates that at lower test-temperatures higher impact values are often obtained with tempering temperatures of 900° F than with 1000° or 1100° F. The curves in Fig. 3 illustrate one of the more pronounced cases. This phenomenon usually occurred in the range of approx. 0° to — 80° F.

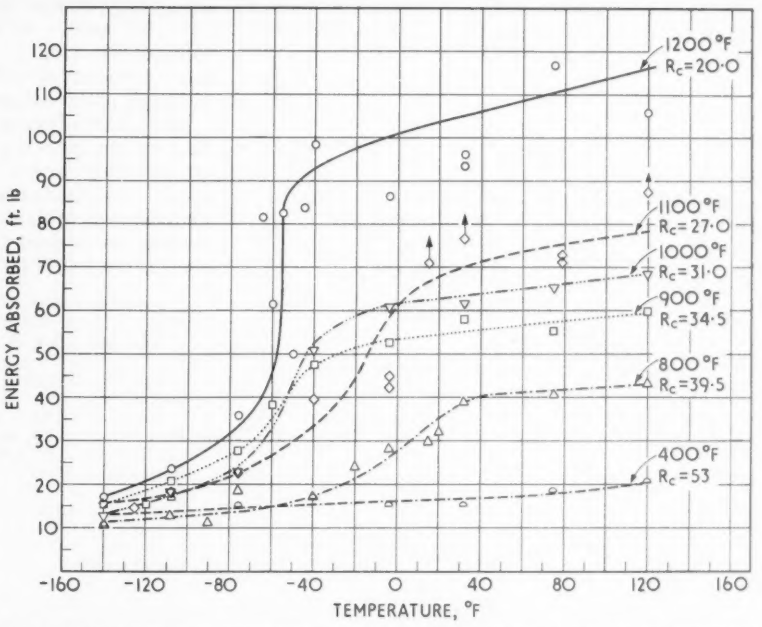

Fig. 3—Impact/test-temperature curves of steel C10 as tempered at indicated temperatures (*Note:* The upward-pointing arrows indicate the values for specimens that failed to break completely); 0·0006% B added as addition agent *B*

Values from all of the impact-energy/test-temperature curves were replotted to evaluate the effect of tempering temperature on impact energy absorbed at several arbitrarily selected test temperatures. The plot for steel C10 is shown in Fig. 4. The phenomenon just discussed generally appeared most pronounced at a test temperature of — 40° F. The practical significance of this is that a steel tempered at 900° F, despite its higher tensile strength, may be used with assurance that it will have less tendency toward brittle fracture under impact loading at sub-zero temperature than the same steel tempered at 1000° or 1100° F to a lower tensile strength. It is possible that this is a manifestation of temper brittleness, but no tests were conducted to evaluate the influence of cooling rate from these tempering temperatures.

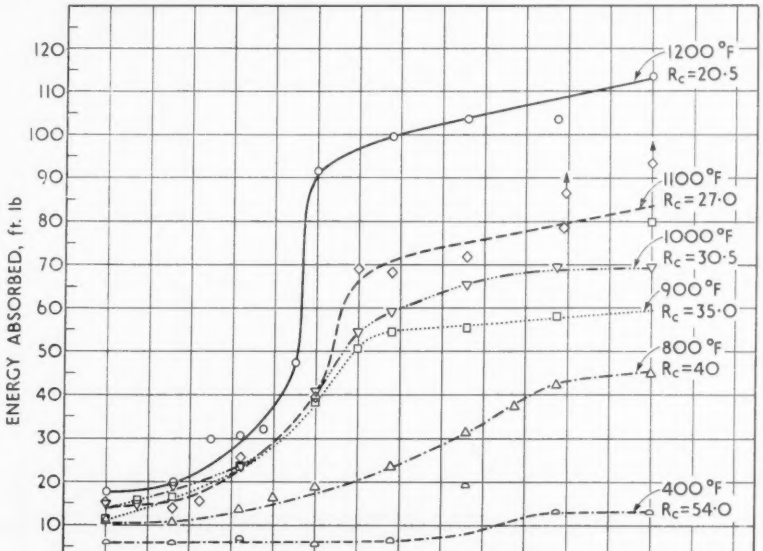

Fig. 2—Impact/test-temperature curves of steel C15 (base steel) tempered at indicated temperatures and air-cooled (*Note:* The upward-pointing arrows indicate the values for specimens that failed to break completely)

As an aid to visualizing the effect of boron on the impact properties of the steels, Fig. 5 was prepared by taking from the various curves the values of the impact energy absorbed at the two arbitrarily selected temperatures of + 70° and — 40° F. These values have been plotted as a function of the boron content for the steels as they were variously tempered. Comparing these values with

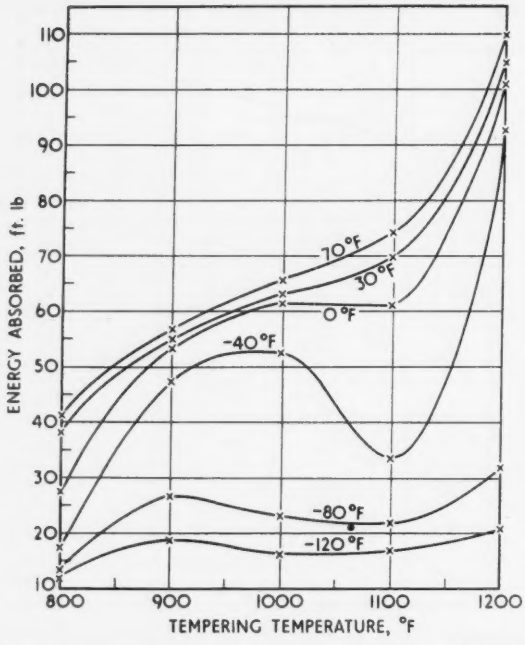

Fig. 4—Impact/tempering-temperature curves of steel C10 at several arbitrary test temperatures, as indicated on curves; 0·0006% B added as addition agent *B*

those for the untreated base steel (broken line in each graph), it is apparent that the addition of boron has a generally adverse effect upon the impact properties at + 70° and − 40°F of the steels tempered in the range of 800–1200° F. This adverse effect is generally more pronounced with the higher contents of boron, although many exceptions to this may be noted.

The effects of boron additions with different intensifiers on the impact properties of the steels tempered at 800–1200° F are shown by the curves in Figs. 6–12. Considering the effect of the amount of boron and the type of intensifier on the impact-energy/test-temperature curves (excluding the temper-embrittled steels), the following observations may be made:

Addition agent A (0·0009, 0·0016, and 0·0031% B, Fig. 6): Markedly detrimental effect.

Addition agent B (0·0003, 0·0006, and 0·0011% B, Fig. 7): Effect not clearly defined, sometimes slightly beneficial, sometimes slightly detrimental. No significant effect.

Addition agent C (0·0022, 0·0028, and 0·0036% B, Fig. 8): Slightly detrimental effect, becoming more marked with increasing amounts of boron for steels tempered at 800°, 900°, and 1000° F. For steels tempered at 1100° and 1200° F, a slightly beneficial effect resulted from the smaller additions and a detrimental effect was exerted only by the largest amount of boron (0·0036%). This addition agent caused a definite coarsening of the austenite grains (*see* Table I).

Addition agent D (0·0012, 0·0025, and 0·0038% B, Fig. 9): Detrimental effect, becoming very marked with increasing amounts of boron.

Addition agent E (0·0001, 0·0004, and 0·0006% B, Fig. 10): Detrimental effect, negligible with 0·0001% B, becoming more marked with increasing amounts of boron.

Addition agent F (0·0007, 0·0008, and 0·0011% B, Fig. 11): No significant effect with 0·0007% B; detrimental effect with larger amounts except for the steel tempered at 1200° F, where detrimental effect was rather slight.

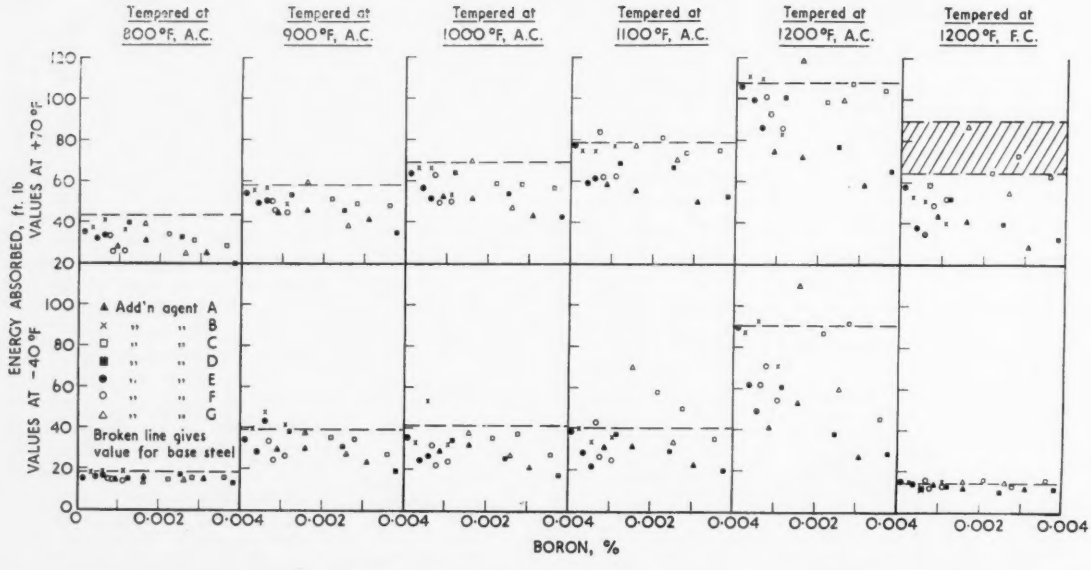

Fig. 5—Effect of boron on the notched-bar impact values of the test steels at + 70° and − 40° F. (The impact-energy/test-temperature curve of the base steel tempered at 1200° F, F.C., was practically vertical at + 70° F, so a definite value could not be selected. The cross-hatched area indicates the upper and lower values for this steel at this temperature)

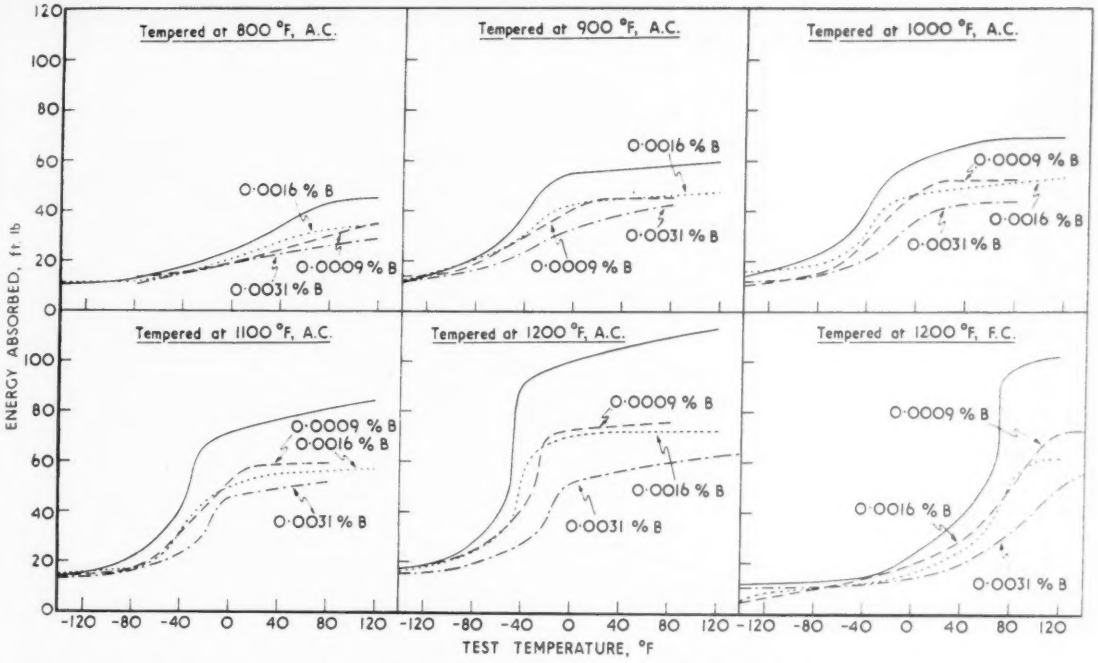
Tempered at 800 °F, A.C.
Tempered at 900 °F, A.C.
Tempered at 1000 °F, A.C.
Tempered at 1100 °F, A.C.
Tempered at 1200 °F, A.C.
Tempered at 1200 °F, F.C.
0·0016 % B
0·0009 % B
0·0031 % B
ENERGY ABSORBED, ft lb
TEST TEMPERATURE, °F

Fig. 6—Effect of boron (added as addition agent *A*) on the impact properties of the steels tempered as indicated. Solid line represents the base steel C15

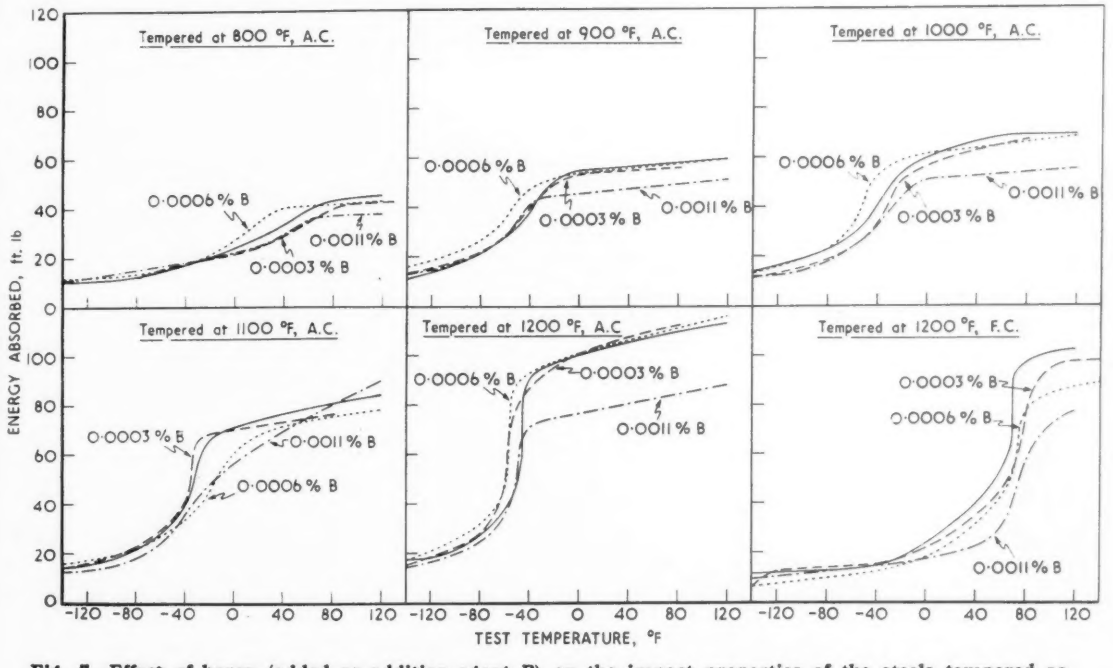
Tempered at 800 °F, A.C.
Tempered at 900 °F, A.C.
Tempered at 1000 °F, A.C.
Tempered at 1100 °F, A.C.
Tempered at 1200 °F, A.C
Tempered at 1200 °F, F.C.
0·0006 % B
0·0011 % B
0·0003 % B
ENERGY ABSORBED, ft. lb
TEST TEMPERATURE, °F

Fig. 7—Effect of boron (added as addition agent *B*) on the impact properties of the steels tempered as indicated. Solid line represents the base steel C15

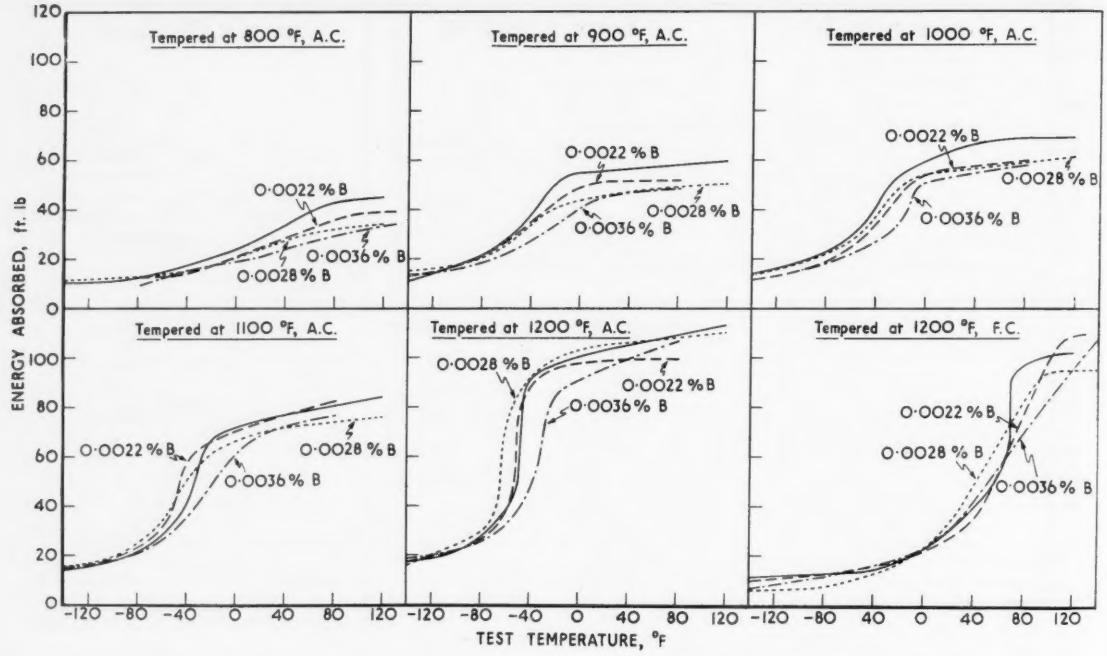
Tempered at 800 °F, A.C.
Tempered at 900 °F, A.C.
Tempered at 1000 °F, A.C.
Tempered at 1100 °F, A.C.
Tempered at 1200 °F, A.C.
Tempered at 1200 °F, F.C.
0·0022 % B
0·0028 % B
0·0036 % B
ENERGY ABSORBED, ft. lb
TEST TEMPERATURE, °F

Fig. 8—Effect of boron (added as addition agent *C*) on the impact properties of the steels tempered as indicated. Solid line represents the base steel C15

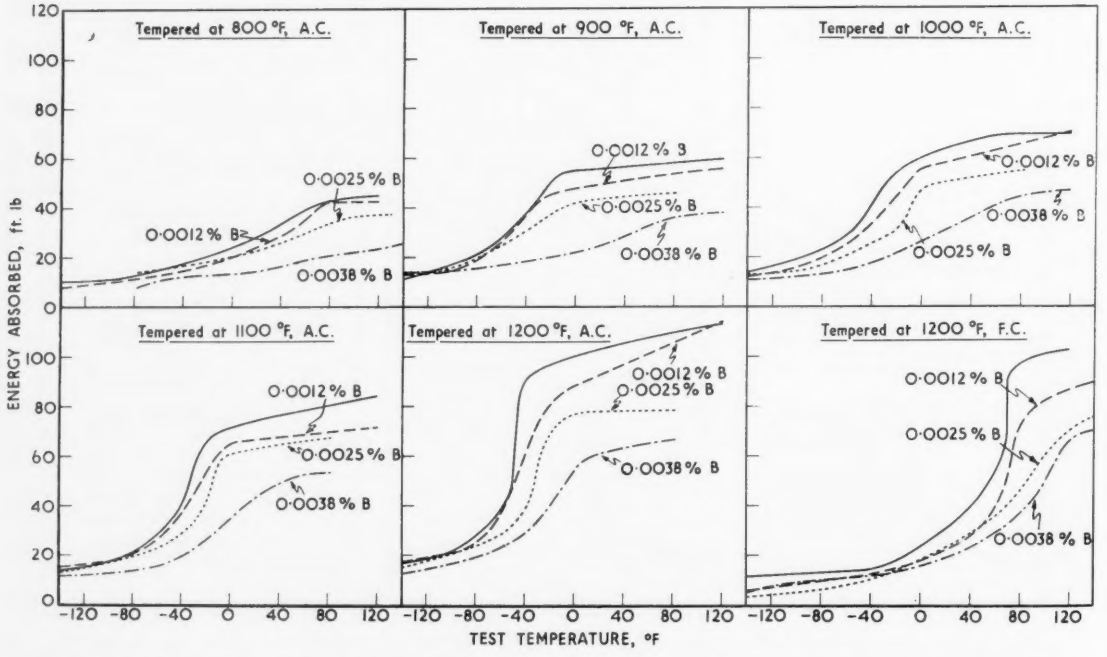
Tempered at 800 °F, A.C.
Tempered at 900 °F, A.C.
Tempered at 1000 °F, A.C.
Tempered at 1100 °F, A.C.
Tempered at 1200 °F, A.C.
Tempered at 1200 °F, F.C.
0·0012 % B
0·0025 % B
0·0038 % B
ENERGY ABSORBED, ft. lb
TEST TEMPERATURE, °F

Fig. 9—Effect of boron (added as addition agent *D*) on the impact properties of the steels tempered as indicated. Solid line represents the base steel C15

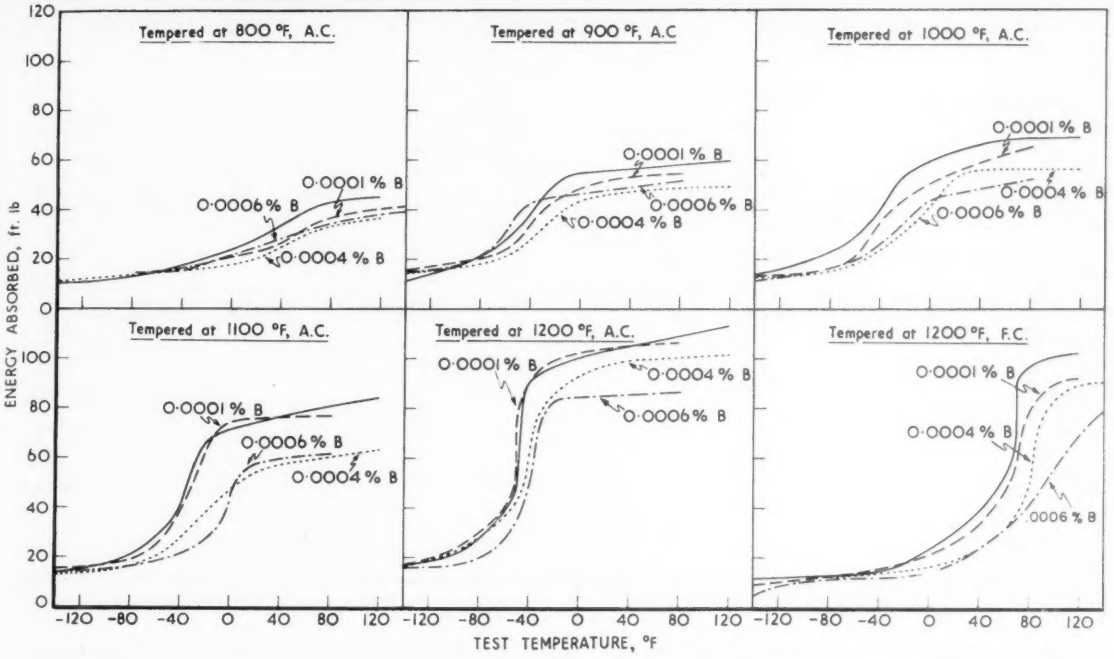
Tempered at 800 °F, A.C.
Tempered at 900 °F, A.C
Tempered at 1000 °F, A.C.
Tempered at 1100 °F, A.C.
Tempered at 1200 °F, A.C.
Tempered at 1200 °F, F.C.
0·0001 % B
0·0006 % B
0·0004 % B
ENERGY ABSORBED, ft. lb
TEST TEMPERATURE, °F

Fig. 10—Effect of boron (added as addition agent *E*) on the impact properties of the steels tempered as indicated. Solid line represents the base steel C15

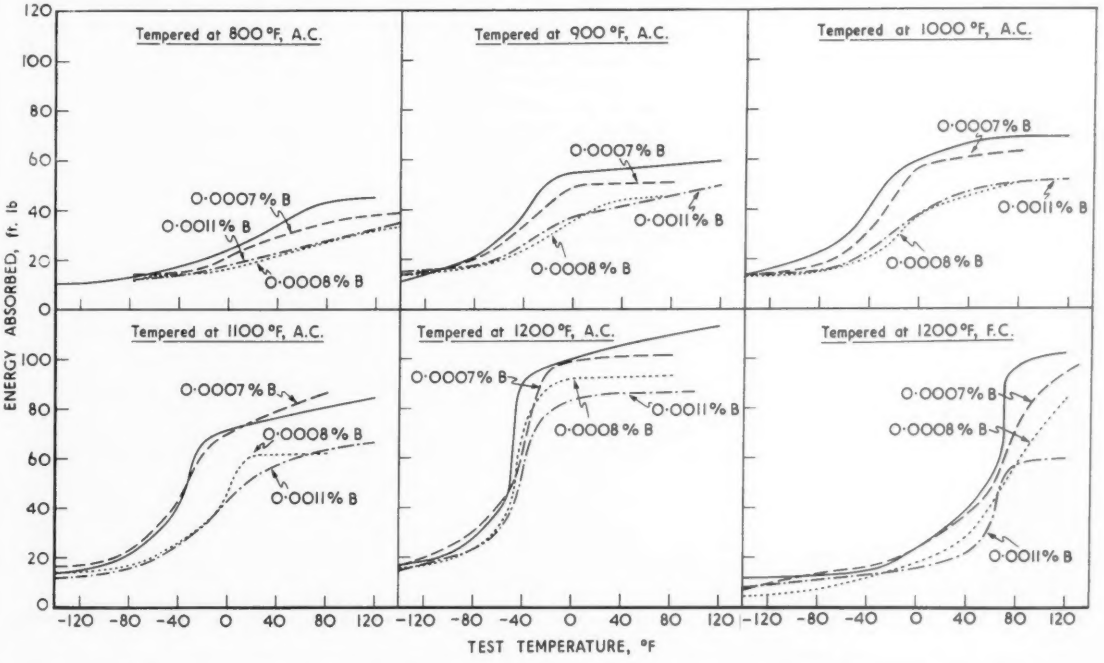
Tempered at 800 °F, A.C.
Tempered at 900 °F, A.C.
Tempered at 1000 °F, A.C.
Tempered at 1100 °F, A.C.
Tempered at 1200 °F, A.C.
Tempered at 1200 °F, F.C.
0·0007 % B
0·0011 % B
0·0008 % B
ENERGY ABSORBED, ft. lb
TEST TEMPERATURE, °F

Fig. 11—Effect of boron (added as addition agent *F*) on the impact properties of the steels tempered as indicated. Solid line represents the base steel C15

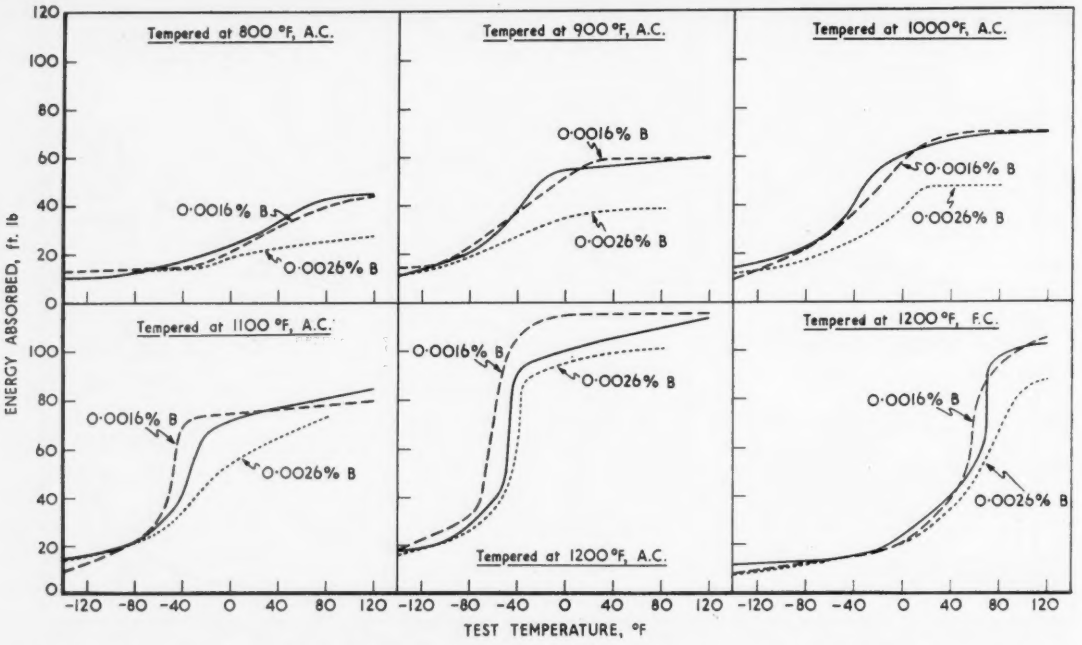
Tempered at 800 °F, A.C.
Tempered at 900 °F, A.C.
Tempered at 1000 °F, A.C.
Tempered at 1100 °F, A.C.
Tempered at 1200 °F, A.C.
Tempered at 1200 °F, F.C.
0·0016% B
0·0026% B
ENERGY ABSORBED, ft. lb
TEST TEMPERATURE, °F

Fig. 12—Effect of boron (added as addition agent *G*) on the impact properties of the steels tempered as indicated. Solid line represents the base steel C15

Addition agent G (0·0016 and 0·0026% B, Fig. 12): With smaller amount, no change except for steels tempered at 1100° and 1200° F, where a moderately beneficial effect was observed; detrimental effect with larger amount except for steel tempered at 1200° F. With this addition agent, 0·0026% B caused a definite coarsening of the austenite grains (*see* Table I).

The transition temperatures of the test steels, determined by the six arbitrarily selected criteria previously described, are plotted as functions of the boron content in Figs. 13–18. Related pertinent data are also plotted in these diagrams. As before, the broken lines represent values for the base steel.

Surveying Figs. 13–18, it is apparent that, regardless of the criterion used, boron generally raises the transition temperatures of the various steels, even though numerous exceptions may be observed. It is also apparent that considerable variation often exists in the relative transition temperature of any particular steel (compared with the base steel) when the different

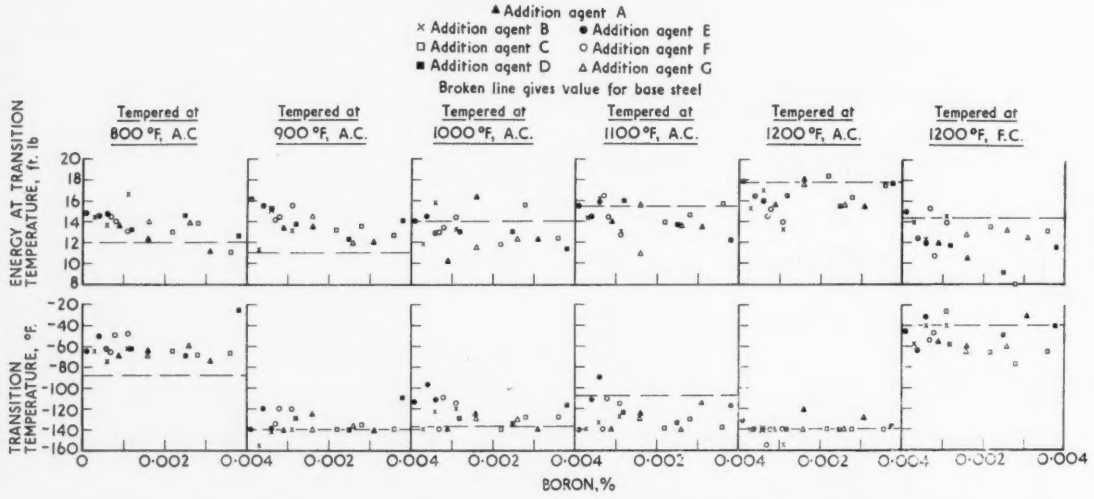
Addition agent A
Addition agent B
Addition agent C
Addition agent D
Addition agent E
Addition agent F
Addition agent G
Broken line gives value for base steel
Tempered at 800 °F, A.C.
Tempered at 900 °F, A.C.
Tempered at 1000 °F, A.C.
Tempered at 1100 °F, A.C.
Tempered at 1200 °F, A.C.
Tempered at 1200 °F, F.C.
ENERGY AT TRANSITION TEMPERATURE, ft. lb
TRANSITION TEMPERATURE, °F
BORON, %

Fig. 13—Chart showing relation between boron content, transition temperature (criterion = 100% brittle fracture), and impact energy absorbed of the test steels tempered as indicated

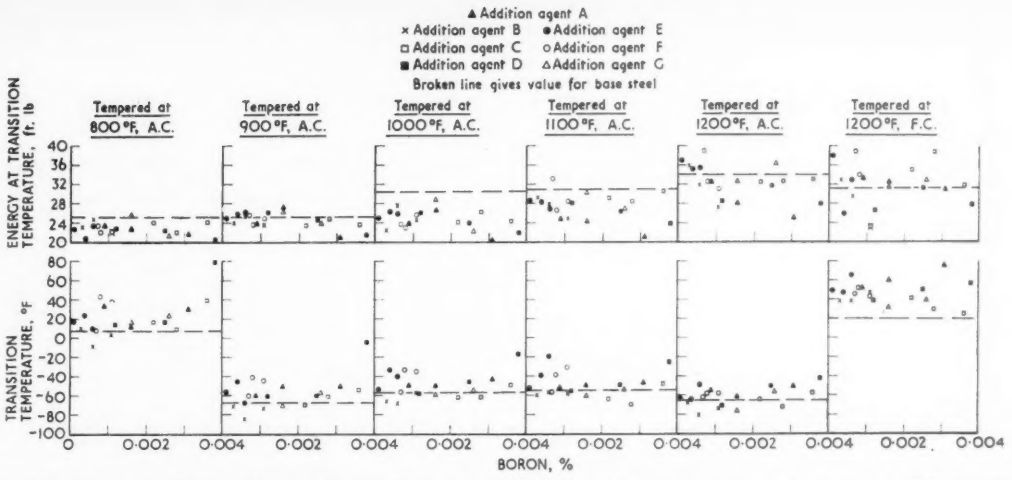
▲ Addition agent A
× Addition agent B
● Addition agent E
□ Addition agent C
○ Addition agent F
■ Addition agent D
△ Addition agent G
Broken line gives value for base steel
Tempered at 800°F, A.C.
Tempered at 900°F, A.C.
Tempered at 1000°F, A.C.
Tempered at 1100°F, A.C.
Tempered at 1200°F, A.C.
Tempered at 1200°F, F.C.
ENERGY AT TRANSITION TEMPERATURE, ft. lb
TRANSITION TEMPERATURE, °F
BORON, %

Fig. 14—Chart showing relation between boron content, transition temperature (criterion = 50% brittle fracture), and impact energy absorbed of the test steels tempered as indicated

criteria are used. Erroneous conclusions may be drawn from observations based solely on arbitrary transition temperatures without reference to the entire impact/test-temperature curves.

Study of these figures reveals some interesting facts. In Fig. 13, using 100% brittle fracture as the criterion of transition temperature, increasing the tempering temperature from 800° to 900° F lowered the transition temperature of most of the boron-treated steels from the range of about — 50° to — 70° F to the range of — 120° to — 140° F. This is a great improvement. Increasing the tempering temperature to 1000° or 1100° F caused only a slight additional change in the transition temperature, and this change was generally adverse, i.e. the transition temperatures were raised a little. This phenomenon emphasizes the observation noted previously that, at low temperatures, the steels tempered at 900° F appeared to have superior notched-bar toughness, despite their higher hardness, than the steels tempered at 1000° and

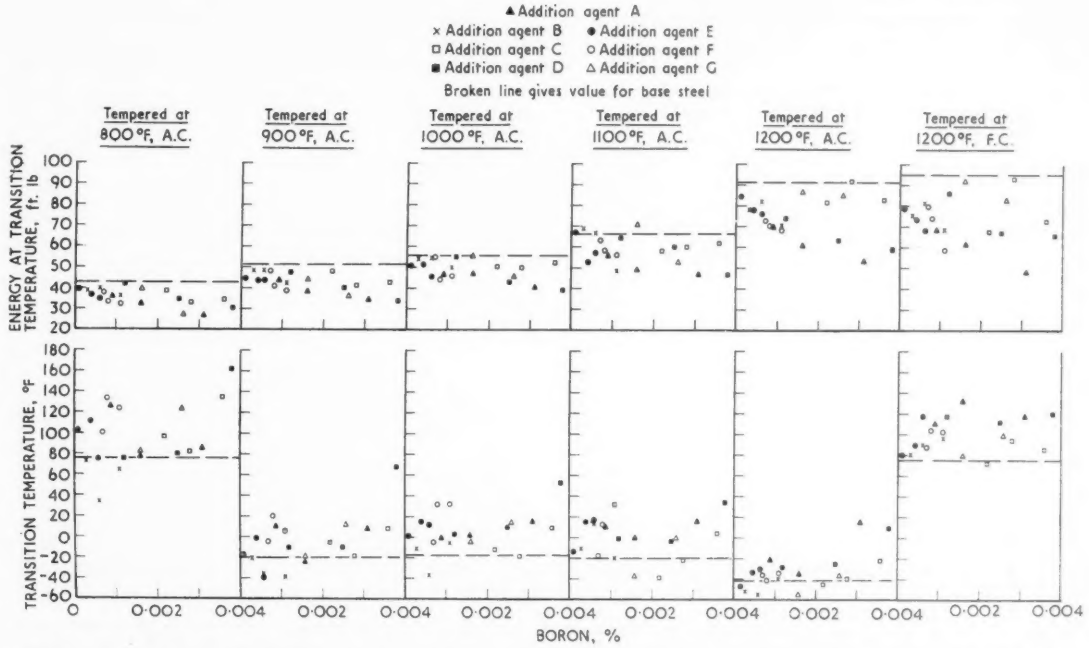
▲ Addition agent A
× Addition agent B
● Addition agent E
□ Addition agent C
○ Addition agent F
■ Addition agent D
△ Addition agent G
Broken line gives value for base steel
Tempered at 800°F, A.C.
Tempered at 900°F, A.C.
Tempered at 1000°F, A.C.
Tempered at 1100°F, A.C.
Tempered at 1200°F, A.C.
Tempered at 1200°F, F.C.
ENERGY AT TRANSITION TEMPERATURE, ft. lb
TRANSITION TEMPERATURE, °F
BORON, %

Fig. 15—Chart showing relation between boron content, transition temperature (criterion = 0% brittle fracture), and impact-energy absorbed of the test steels tempered as indicated

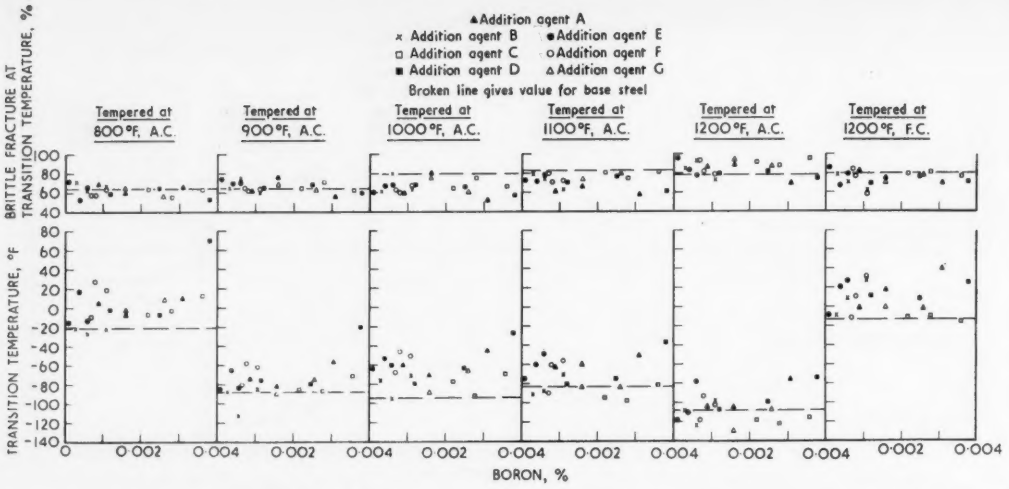
▲Addition agent A
× Addition agent B
● Addition agent E
□ Addition agent C
○ Addition agent F
■ Addition agent D
△ Addition agent G
Broken line gives value for base steel
Tempered at 800°F, A.C.
Tempered at 900°F, A.C.
Tempered at 1000°F, A.C.
Tempered at 1100°F, A.C.
Tempered at 1200°F, A.C.
Tempered at 1200°F, F.C.
BRITTLE FRACTURE AT TRANSITION TEMPERATURE, %
TRANSITION TEMPERATURE, °F
BORON, %

Fig. 16—Chart showing relation between boron content, transition temperature (criterion = 20 ft-lb), and per cent brittle fracture of the test steels tempered as indicated

1100° F. The same general trend is noticeable in Figs. 14–18. It is surprising that the transition temperatures of the steels tempered at 1200° F are, generally, only slightly lower than those tempered at 1100° F.

With the transition temperature criterion of 100% brittle fracture (Fig. 13), it appeared that the energy absorbed was relatively insensitive to tempering temperature. With the criteria of 50% and 0% brittle fracture (Figs. 14 and 15), however, there was a definite trend toward higher impact values with increasing tempering temperatures. This same trend was evident with the criterion of $\frac{1}{2}(E_{100\%\ B} + E_{0\%\ B})$ (Fig. 18).

Effect of Boron on Temper Brittleness

Curves showing the effect of air *v.* furnace cooling (at the selected rate of 30° F/h) on the impact properties and fracture characteristics as a function of test temperature were plotted for all of the steels tested. The curves for the base steel (C15) are shown in Fig. 19. All the steels were susceptible to temper embrittlement, as was observed by the displacement of the impact-test temperature and fracture-test temperature curves to higher test temperatures by the embrittling treatment.

The cross-hatched area in Fig. 19 was constructed by joining the points on the impact curves (of the steels as air-cooled and as furnace-cooled) correspond-

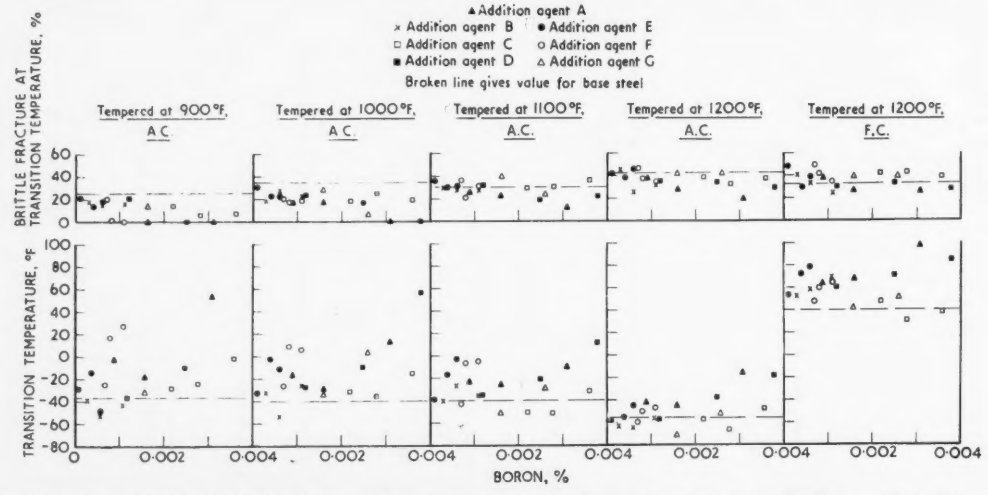
▲Addition agent A
× Addition agent B
● Addition agent E
□ Addition agent C
○ Addition agent F
■ Addition agent D
△ Addition agent G
Broken line gives value for base steel
Tempered at 900°F, A.C.
Tempered at 1000°F, A.C.
Tempered at 1100°F, A.C.
Tempered at 1200°F, A.C.
Tempered at 1200°F, F.C.
BRITTLE FRACTURE AT TRANSITION TEMPERATURE, %
TRANSITION TEMPERATURE, °F
BORON, %

Fig. 17—Chart showing relation between boron content, transition temperature (criterion = 40 ft-lb), and per cent brittle fracture of the test steels tempered as indicated. (No graph is shown for the steels tempered at 800° F, as in this condition only a few steels absorbed as much as 40 ft-lb. Some did not absorb this amount of energy even as tempered at 900° F)

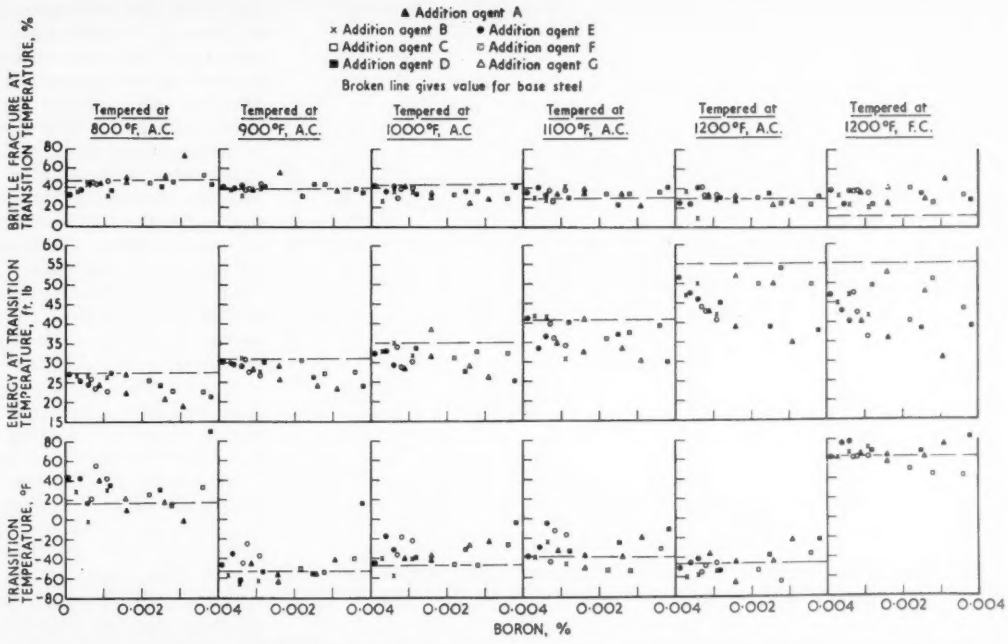
▲ Addition agent A
× Addition agent B
● Addition agent E
□ Addition agent C
▫ Addition agent F
■ Addition agent D
△ Addition agent G
Broken line gives value for base steel
Tempered at 800°F, A.C.
Tempered at 900°F, A.C.
Tempered at 1000°F, A.C.
Tempered at 1100°F, A.C.
Tempered at 1200°F, A.C.
Tempered at 1200°F, F.C.
BRITTLE FRACTURE AT TRANSITION TEMPERATURE, %
ENERGY AT TRANSITION TEMPERATURE, ft. lb
TRANSITION TEMPERATURE, °F
BORON, %

Fig. 18—Chart showing relation between boron content, transition temperature (criterion = average breaking energy between 100% and 0% brittle fracture), impact energy absorbed, and per cent brittle fracture of the steels tempered as indicated

ing to 100% and 0% brittle fracture, as described by Taber, Thorlin, and Wallace.[9] The average temperature shift (marked 'A.T.S.' in the diagram) was calculated for each steel by dividing the cross-hatched area (measured with a planimeter) by the average difference in ft-lb. The value thus obtained is a quantitative measure of the shift (in ° F) of the impact curve caused by the embrittling treatment.

The effects of boron additions with different intensifiers on the impact properties of the steels as tempered at 1200° F and furnace-cooled are shown by the appropriate curves in Figs. 6–12. Considering the effect of the amount of boron and the type of intensifier on the impact-energy/test-temperature curves, the following observations may be made:

Addition agent A (0·0009, 0·0016, and 0·0031% B, Fig. 6): Detrimental effect, becoming more marked with increasing amounts of boron.

Addition agent B (0·0003, 0·0006, and 0·0011% B, Fig. 7): Detrimental effect, negligible with 0·0003% B, becoming more marked with 0·0011% B.

Addition agent C (0·0022, 0·0028, and 0·0036% B, Fig. 8): Negligible effect. This addition agent caused a definite coarsening of the austenite grains (*see* Table I).

Addition agent D (0·0012, 0·0025, and 0·0038% B, Fig. 9): Detrimental effect, becoming more marked with increasing amounts of boron.

Addition agent E (0·0001, 0·0004, and 0·0006% B, Fig. 10): Detrimental effect, negligible with 0·0001% B, becoming more marked with increasing amounts of boron.

Addition agent F (0·0007, 0·0008, and 0·0011% B, Fig. 11): Detrimental effect, negligible with 0·0007% B, becoming more marked with 0·0011% B.

Addition agent G (0·0016 and 0·0026% B, Fig. 12): Slightly detrimental effect with 0·0026% B. The austenite grains were coarsened with this addition (*see* Table I).

This rather general evaluation, it will be observed, does not quite agree with the evaluation previously given of the effect of the various addition agents on the notched-bar toughness of the steels tempered at the various temperatures and air-cooled.

These curves indicate that the temper brittleness present in the base steel is aggravated when boron is added in any form other than as simple intensifiers (addition agents *C* and *G*). It is significant that the other five addition agents contained titanium and aluminium in addition to boron; three of them contained high percentages of silicon, three contained zirconium, two vanadium, and one each calcium and magnesium. As all of the steels contained high manganese and appreciable amounts of silicon and aluminium, the presence of these elements in the addition agents may be ruled out as the cause of increased temper brittleness. The one element common to these five addition agents, and that which appeared to be responsible for the detrimental effects observed, is therefore titanium. From the chemical similarity of zirconium and titanium, it may be expected that zirconium might have the same effect as titanium. Although none of the intensifiers contained vanadium or zirconium in the absence of titanium, the presence of vanadium or zirconium did not reduce the adverse effects apparently caused by titanium. It is clear that neither silicon nor boron is to be blamed. The observed increase in susceptibility to temper brittleness of many of the boron-treated steels may therefore be ascribed to the presence of other elements introduced with the complex intensifiers and not to the element boron.

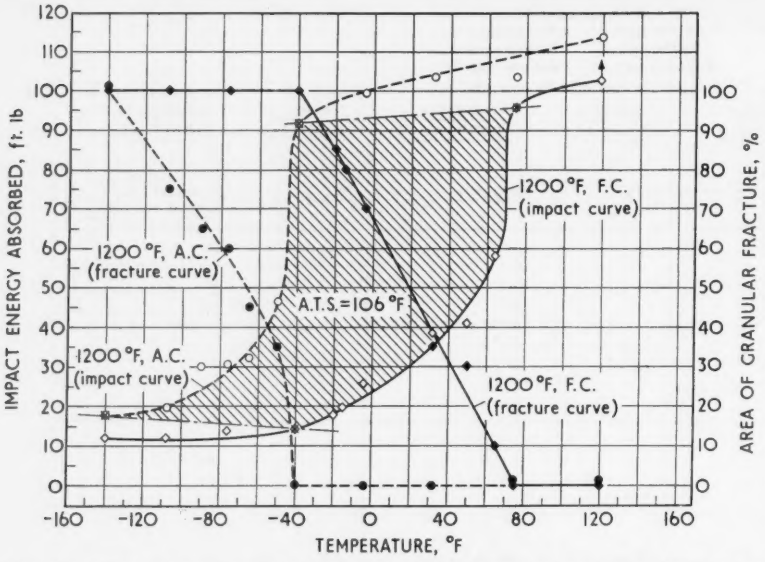
IMPACT ENERGY ABSORBED, ft. lb
AREA OF GRANULAR FRACTURE, %
TEMPERATURE, °F
1200 °F, A.C. (fracture curve)
1200 °F, A.C. (impact curve)
A.T.S.=106 °F
1200 °F, F.C. (impact curve)
1200 °F, F.C. (fracture curve)

Fig. 19—Effect of air cooling *v.* furnace-cooling (after tempering at 1200° F) on the impact properties of steel C15 (base steel, no B added)

The transition temperatures of the air-cooled and furnace-cooled steels, selected according to the six criteria previously described, are given in Figs. 13–18. Although there is considerable scatter in the data, one general trend is apparent in all these diagrams. When air-cooled from 1200° F, the transition temperatures of the boron-treated steels fall both above and below that of the base steel in such a fashion as to indicate that boron has but little definite influence on the transition temperatures of the steels treated in this way. This is different from the observation previously noted for the general behaviour of the steels as tempered over the entire range of 800–1200° F and air-cooled. However, the transition temperatures of most of the boron-treated steels as furnace-cooled from 1200° F are higher than those of the base steel, indicating that the addition agents generally increase the susceptibility to temper brittleness. The values that most frequently fall in the vicinity of the base steel correspond to the addition agents *C* and *G* (open triangles and squares), which additions have been shown previously to have negligible effect upon the energy-absorbed/test-temperature curves of the temper-embrittled steels.

Even though some correlation appears to exist between boron content, energy absorbed at transition temperature, or percentage of brittle fracture, the trends were not definite enough to be significant.

In Fig. 20 the shift in transition temperature is plotted as a function of boron content for each criterion of transition temperature. Included also is a graph for the 'A.T.S.' values (area measurements) previously described. This shift is the arithmetic difference between the transition temperatures of the furnace-cooled and air-cooled steels. The smaller the value, therefore, the less is the temper embrittlement.

Since the numerical value of the shift in transition temperature takes into account the transition temperature of the steel both as air-cooled and as furnace-cooled, it would seem that this is a logical method for evaluating the embrittling effect of the various boron additions. This method of measurement does not, however, take into account the possible deleterious effect of the particular boron addition on the transition temperature of the air-cooled steel. To illustrate, consider the effect of 0·0031% B added as addition agent *A* (solid triangle), using the criterion of 0% brittle fracture. In Fig. 20, this point falls well below the value for the base steel, indicating that the tendency toward temper brittleness is reduced when compared with the base steel. In Fig. 15, however, the transition temperature of the furnace-cooled boron-treated steel is about 40° F higher than that of the base steel similarly treated, a very definite indication that this boron addition increases the tendency toward temper embrittlement. The apparently beneficial effect of the boron addition shown in Fig. 20 is a reflection of the injurious effect on the transition temperature of the air-cooled steel (Fig. 15). Here the boron addition raises the transition temperature 60° F over that of the base steel.

The summary curves in Fig. 20 may thus be used only to evaluate the effect of the various boron additions on the tendency to temper embrittlement without giving any indication whether any particular addition raises the transition temperature of the air-cooled steel so that the additional increase caused by furnace-cooling appears less injurious.

Whether or not any particular boron-treated steel appears to be more or less sensitive to temper embrittlement depends on the criterion of measurement used. It is evident from a study of Fig. 20 that the steels are often rated quite differently by the various criteria. It must be emphasized that, in order to evaluate temper brittleness, the entire energy-absorbed/test-temperature curve should be considered.

CONCLUSIONS

A series of 21 steels, made to a base composition of 0·43% carbon and 1·60% manganese, were studied to evaluate the effect of boron on the impact properties over the temperature range − 140° to + 140° F. All steels were from the same open-hearth heat and the boron additions were made in the mould by means of seven different intensifiers. Charpy impact tests were made on **V**-notched specimens as fully hardened and then tempered at 400°, 800°, 900°, 1000°, 1100°, and 1200° F, followed by air-cooling. To study temper brittleness, the treatment utilized was tempering at 1200° F, followed by furnace-cooling at the rate of 30° F/h. It was concluded that:

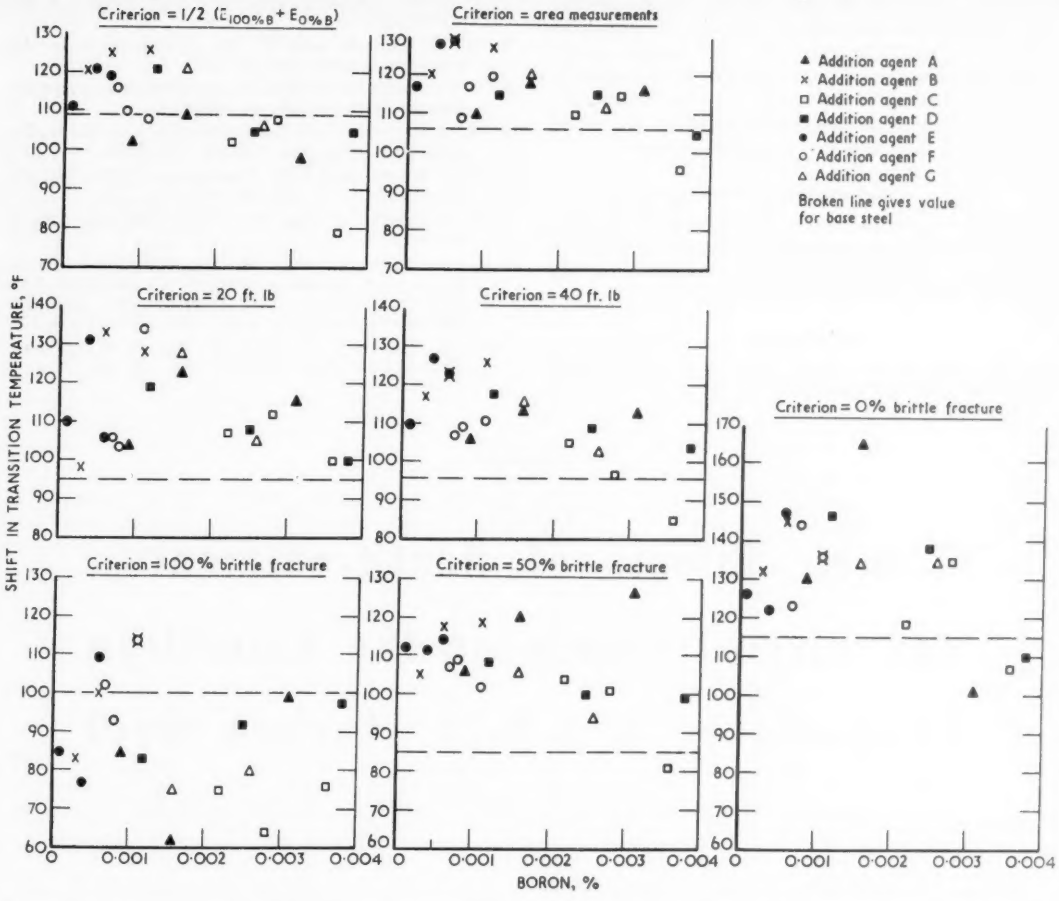
Criterion = 1/2 ($E_{100\%B}$ + $E_{0\%B}$)
Criterion = area measurements
Addition agent A
Addition agent B
Addition agent C
Addition agent D
Addition agent E
Addition agent F
Addition agent G
Broken line gives value for base steel
Criterion = 20 ft. lb
Criterion = 40 ft. lb
Criterion = 0% brittle fracture
Criterion = 100% brittle fracture
Criterion = 50% brittle fracture
SHIFT IN TRANSITION TEMPERATURE, °F
BORON, %

Fig. 20—Summary chart showing effect of boron on the shift in transition temperature caused by embrittling treatment as determined by various criteria

(1) The notched-bar impact values of the boron-treated steels as tempered at 400° F were superior to those of the untreated steel.

(2) The notched-bar impact values of the boron-treated steels as tempered at 800°, 900°, 1000°, 1100°, and 1200° F (air-cooled) were generally inferior to those of the untreated steel, although many exceptions were noted. The extent of this detrimental effect depended upon the type of addition agent, the amount of boron, and the tempering temperature at which the steels had been treated. As a general statement, the detrimental effect can be minimized or avoided by the use of addition agent *B* up to 0·0011% B (larger amounts of boron were not evaluated); addition agent *C* up to 0·0028% B; addition agent *F* up to 0·0007% B; and addition agent *G* up to 0·0016% B.

(3) The steels tempered at 900° F had superior notched-bar toughness at sub-zero temperatures, despite their higher hardness, than the steels tempered at 1000° or 1100° F. This was quite evident in the temperature range 0° to — 80° F and usually was most marked in the neighbourhood of — 40° F. The practical significance of this is that the stronger steels, tempered at 900° F, may be used with assurance that they will have less tendency toward brittle fracture under impact loading at sub-zero temperature than the weaker steels, tempered at 1000° or 1100° F.

(4) Temper brittleness of the base steel was increased when boron was added by means of the complex intensifiers (addition agents *A*, *B*, *D*, *E*, and *F*). When added in the form of simple intensifiers (addition agents *C* and *G*), boron had a negligible effect on temper brittleness. The increased susceptibility to temper embrittlement of many of the boron-treated steels may therefore be ascribed to certain other elements that were introduced by the addition agents and not to boron itself. It is believed that the element responsible is titanium or its sister element zirconium.

(5) Considering the effect of boron on the steels as fully hardened and tempered between 800° and 1200° F (air-cooled) and also as temper embrittled (furnace-cooled from 1200° F), it appears that boron may be added in amounts up to about 0·0025% in the form of simple intensifiers (addition agents *C* or

G) without adversely affecting the impact properties.

(6) Evaluation by various criteria of transition temperatures from the many curves of impact-energy *v.* test-temperature showed many inconsistencies. There is a great need for a precise, understandable, and standardized method of evaluating transition temperatures from notched-bar impact data.

Acknowledgments

The authors are indebted to Mr. Thomas P. Royston for his assistance in testing the steels and making the illustrations. The investigation was sponsored by the Ordnance Corps, Watertown Arsenal.

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Some Effects of Heat-treatment and Microstructure on the Transition Temperature of a 0·24% Carbon Steel

By G. Burns, M.Met., F.I.M., and C. Judge, A.Met., A.I.M.

SYNOPSIS

The relationships between transition temperature in a notched bend test, microstructure, and thermal history have been examined in the case of a ½-in. thick plate of 0·24% carbon mild steel.

Austenite grain size at the commencement of cooling is shown to be a major factor affecting the transition temperature. Cooling rate also has a substantial effect which is more marked during the transformation of the pearlite than in other parts of the cooling range.

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AMONG THE FACTORS known to affect the temperature at which the fracture of mild steel in a notched-bend test changes from a tough to a brittle character is the thermal history of the material, particularly that part of the history which follows the conclusion of hot-working. If, as is usually the case, the material is supplied and used in the 'as rolled' or the normalized condition, the thermal history can be defined in terms of either the temperature at which rolling finished or at which normalizing was carried out, and the cooling rate from that temperature.

These factors of the thermal history also affect the microstructure of the steel, and it seems reasonable to seek some correlation between microstructure and transition temperature. One of the advantages of such a correlation, if it were found, would lie in the fact that different features of the microstructure are established in different parts of the cooling range and the correlation could show which portions of the cooling range are more, and which less, significant from the point of view of transition temperature.

Some indications of correlations between transition temperature of mild steels and microstructure have been mentioned by several workers. Barr and Tipper[1] and MacKenzie[2] have noted a correlation between ferrite grain size and transition temperature, whilst Petch[3] has shown a relationship between ferrite grain size and brittle strength, a property which may be expected to influence the transition temperature. Grossman[4] investigated the relationship of transition temperature to coarseness of the pearlite and found a fine pearlite structure to be associated with a low transition temperature. Barr and Honeyman[5] observed indications of correlation between austenite grain size and transition temperature, whilst MacKenzie,[2] examining 'as rolled' plates, showed a correlation between transition and finishing temperatures, a factor affecting the austenite grain size.

In some cases these indications of relationships have been derived from the examination of samples coming

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from numerous casts of mild steel among which there must inevitably be variations, though generally small ones, of such factors as composition, melting history, and hot-working, which are known to affect the transition temperature. Though it is unlikely that the small and random variations of these additional factors have substantially affected the conclusions drawn, an attempt has been made in the present work to limit as far as possible the potential disturbing factors by confining the work to material from one plate only.

MATERIAL

The steel used throughout the work was a ½-in. thick plate of composition:

Carbon, %	0·24
Silicon, %	0·15
Manganese, %	0·55
Sulphur, %	0·038
Phosphorus, %	0·033
Nickel, %	0·15
Chromium, %	0·04
Copper, %	0·16

The austenite grain growth characteristics of the steel, determined by heating for 40 min at each temperature and cooling at such a rate as to make the austenite grain size observable in the final microstructure were:

Temperature, ° C	A.S.T.M. Grain Size
850	7
890	6
940	4
1030	4
1090	3
1150	1–2

The times taken for complete isothermal transformation of austenite in the steel at various temperatures in the pearlite range were determined as follows:

Temperature, ° C	Time for Complete Transformation, min
680	10·5
650	2
600	1·5
550	1

During the course of the work, samples have been cooled under several different conditions; the cooling rates over the range 860–600° C corresponding to these conditions and the temperatures of A_3 and A_1 (max.) at these cooling rates, determined in a simple dilatometer, were:

Condition	Average Cooling Rate 860–600° C, °C/min	A_3, ° C	A_1 (max.), ° C
Furnace *A*	1·25	780	715
Furnace *B*	5	775	683
Still air	70	755	662
Air blast	200	755	659

The cooling rate measured in this instance for cooling in still air may be compared with the cooling rate shown by MacKenzie[2] for ½-in. thick plates cooling on a plate mill floor which, over the same temperature range, is about 65 °C/min.

TESTING PROCEDURE

All heat treatments were carried out on samples of plate of sufficient size to produce a number of test pieces. After treatment the samples were cut to give test pieces 56 mm × 10 mm × 10 mm, with the length in the main direction of rolling of the plate, notched in one of the faces perpendicular to the surface of the plate, using a standard Izod notch (45° **V** notch, 2 mm deep, 0·25 mm root radius).

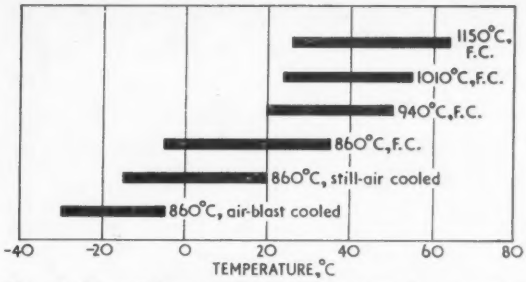


Fig. 1—Transition temperature ranges for samples cooled from varying temperatures and at varying rates

The test pieces were subjected to slow bending in a jig made to fit into the Tensometer machine and giving a maximum angle of bend of 90°. The jig and test piece were enclosed in a thermally insulated box through which hot or cold liquid could be circulated to control the test temperature. Above room temperature the liquid used was ethylene glycol, whilst for low temperatures methylated spirit was used.

It was found that appearance of the fracture surface gave a more definite indication of the inception of the change from tough to brittle behaviour than did the energy absorbed in the test, and throughout the work the transition temperature range has been taken as the range over which the fracture appearance changed from 5% brittle to 95% brittle.

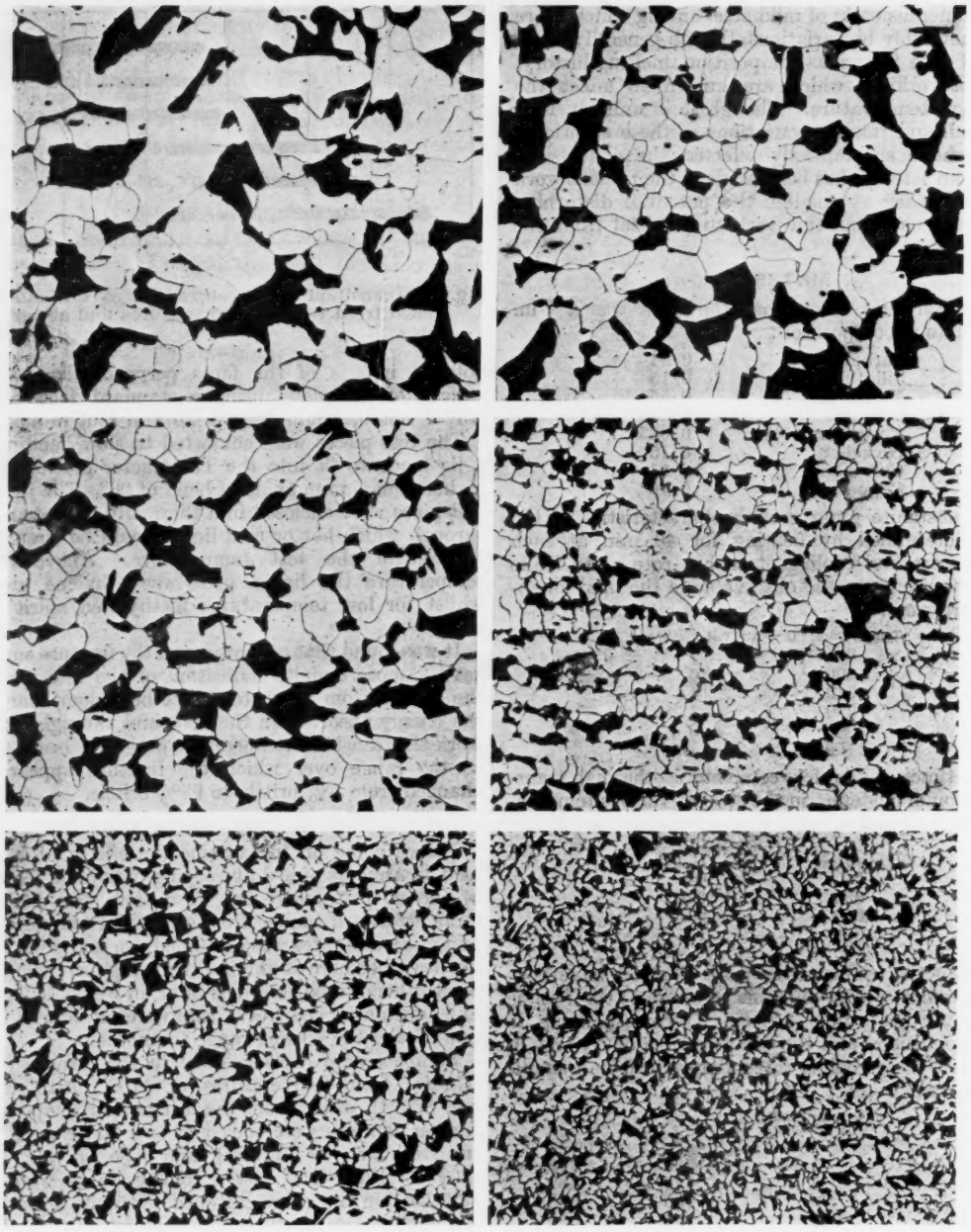
SAMPLES UNIFORMLY COOLED

Before going on to examine the effects of special heat treatments aimed at varying individual features of the microstructure one at a time, the effect of straightforward cooling from a range of temperatures and at several rates was examined; these cooling rates are described here as uniform to distinguish them from some others, to be described later, in which deliberate and substantial changes of rate were introduced from one part of the cooling range to another.

Samples were cooled slowly in the furnace (1·25 °C/min) from temperatures of 1150°, 1010°, 940°, and 860° C, whilst from the last named temperature further samples were cooled at two faster rates, one in still air (70 °C/min) and one in air blast (200 °C/min). The faster cooling rates were not suitable for the coarse grained austenite formed at the higher temperatures since there was relatively little separation of ferrite and the structures obtained could not be described as typical of normal mild steel.

The results of tests carried out on these samples are shown diagrammatically in Fig. 1. In this investigation it has been preferred to consider the upper temperature limit of the transition temperature range as the value to use in making comparisons between samples, and it is on this basis that the results are discussed.

For the samples all furnace cooled from a range of temperatures it will be seen that the temperature from which the heat treatment is carried out has a great



(*a*) F.C. from 1150° C
(*b*) F.C. from 1010° C
(*c*) F.C. from 940° C
(*d*) F.C. from 860° C
(*e*) Cooled in still air from 860° C
(*f*) Cooled in air blast from 860° C

Fig. 2—Structure of steel after different heat-treatments × 125

effect on the transition temperature; the effect appears to be one of austenite grain size principally rather than of temperature *per se* since the transition temperature is raised substantially more by changes of heat treatment temperature from 860° C to 940° C and from 1010° C to 1150° C, both of which correspond to coarsening of the austenite, than by a change of heat treatment temperature from 940° C to 1010° C, over which range there was no perceptible increase in the austenite grain size.

The samples cooled at different rates from 860° C show, within the range of cooling rates used, a progressive and substantial lowering of the transition temperature as the cooling rate is increased.

Fig. 3—Structure of steel after (*a*) and (*b*) treatment *A*, (*c*) treatment *A′* (*a*) and (*c*) × 125, (*b*) × 750

The microstructures of the steel after these six heat treatments are shown in Figs. 2*a–f* where it will be seen that as the heat treatment temperature is lowered and as the cooling rate is increased the ferrite becomes finer and the ferrite–pearlite distribution alters somewhat. Examination at higher magnifications showed, as would be expected, that in the samples cooled at the same rate the pearlite was of about the same degree of coarseness in all, whilst in the samples cooled at different rates the pearlite became finer as the cooling rate increased.

SAMPLES SPECIALLY HEAT-TREATED

In planning special heat-treatments with the object of varying individual features of the microstructure one at a time, it was considered that the initial austenite grain size and the cooling rate down to about 700° C were the factors which would control the ferrite grain size and the size and distribution of the pearlite areas, whilst the cooling rate from about 680° C to 550° C, or, alternatively, temperature of isothermal transformation within this range, would control the coarseness of the pearlite. The cooling rate after completion of the pearlite transformation could be expected to influence the condition of any constituents still in solution at the end of the pearlite transformation, but which might precipitate at lower temperatures.

To ensure, so far as possible, that samples cooled nominally at the same rate in significant regions of the cooling range should, in fact, cool at closely comparable rates, it was thought best to start the significant stages of cooling always from one temperature; for this purpose 860° C, the lowest austenitizing temperature used, was chosen and samples heated to higher initial temperatures to develop coarse austenite grain sizes were transferred to a furnace at 860° C and allowed to settle at this temperature before starting the cooling proper.

Based on these considerations, the following nine heat-treatments were applied to samples of the steel; references are given to the photomicrographs illustrating the structure of the treated steel.

Treatment	Details	Fig. No.
A	Heated at 860° C, furnace cooled to 710° C, quenched to 680° C, held 12 min, cooled in still air.	3*a*, 3*b*
A′	Same as *A*.	3*c*
B	Heated at 860° C, furnace cooled to 710° C, quenched to 550° C, held 2 min, cooled in still air.	4*a*, 4*b*
C	Heated at 860° C, quenched to 710° C, held 2 min, quenched to 680° C, held 12 min, cooled in still air.	5
D	Heated at 860° C, furnace cooled to 710° C, quenched to 680° C, held 12 min, cooled in air blast.	
E	Heated at 860° C, furnace cooled to 710° C, quenched to 550° C, held 2 min, cooled in air blast.	
F	Heated at 860° C, furnace cooled to 710° C, quenched to 550° C, held 2 min quenched in oil.	
G	Heated at 1150° C, transferred to furnace at 860° C, furnace cooled to 710° C, quenched to 680° C, held 12 min, cooled in still air.	6*a*, 6*b*
G′	Same as *G*.	6*c*
H	Heated at 1150° C, transferred to furnace at 860° C, furnace cooled to 710° C, quenched to 550° C, held 2 min, cooled in still air.	7*a*, 7*b*
J	Heated at 1150° C, transferred to furnace at 860° C, quenched to 710° C, held 2 min, quenched to 680° C, held 12 min, cooled in still air.	8

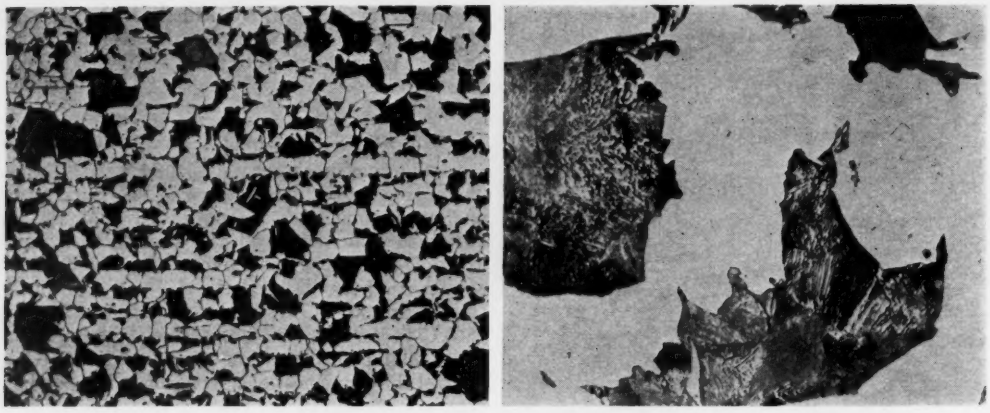


Fig. 4—Structure of steel after treatment *B* (*a*) × 125, (*b*) × 750

In the case of treatment *D*, which differed from treatment *A* only in respect of cooling rate from 680° C, and of treatments *E* and *F*, which differed from treatment *B* only in rate of cooling from 550° C, photomicrographs are not shown as they were indistinguishable from those given treatments *A* and *B*.

Considering these treatments and ignoring, for the moment, the microstructures actually produced, it would seem likely that the following comparisons could be made:

	Treatments
Effect of ferrite grain size	*A* and *C* *G* and *J*
Effect of pearlite structure	*A* and *B* *G* and *H*
Effect of cooling rate below the pearlite transformation range	*A* and *D* *B*, *E*, and *F*
Effect of austenite grain size	*A* and *G* *B* and *H*

Examination of the microstructures indicates that the comparison of *A* and *C* should be valid, as also should that of *G* and *J*, although the differences in ferrite grain size within these two pairs of samples are not very great. Of the comparisons for the effect of pearlite structure, that between samples *A* and *B* should be satisfactory, but the samples given treatments *G* and *H* do not afford so good a comparison since the ferrite–pearlite distribution in these two samples is markedly different.

Bearing in mind the point, mentioned above, that the structure after treatment *D* was similar to that after treatment *A*, whilst treatments *E* and *F* resulted in structures similar to that after treatment *B*, comparisons of these samples for the effect of cooling rate below the pearlite transformation range can be made.

The effects of austenite grain size alone, separated from its effects on other features such as ferrite grain size and ferrite–pearlite distribution cannot be determined from the samples given treatments *A* and *G*, or *B* and *H*, respectively. A number of attempts have been made to produce more comparable final structures from fine and coarse grained austenites, but all have been unsuccessful.

The results of tests on samples given treatments *A* to *J* are shown diagrammatically in Fig. 9.

Considering first the results obtained in the two instances in which batches of material were heat-treated in duplicate (treatments *A*, *A'*, and G, G') although the extent of the transition range varies between the nominally identical batches for each treatment, the temperature at which cleavage fracture begins to appear is identical for each batch given the same treatment, and it seems that the best basis for comparison between treatments would be to take the upper temperature limit of the transition range as the significant value.

Comparison of the results obtained after treatments *A* and *C* shows a lowering of the transition temperature with reduction of ferrite grain size, but comparison of results from treatments *G* and *J* shows a small rise of transition temperature with the finer ferrite grain size. As has already been mentioned, the difference in ferrite grain size between these pairs of samples is not great, and the results are not, therefore, conclusive in establishing the effect which could be expected from large changes of ferrite grain size. In all these samples the ferrite grain size is about equal to, or coarser than, that obtained by furnace cooling from 860° C in the uniform cooling experiments (Fig. 2*d*), and is substantially coarser than would be obtained in practice by free cooling of a medium thickness plate from a proper normalizing temperature.

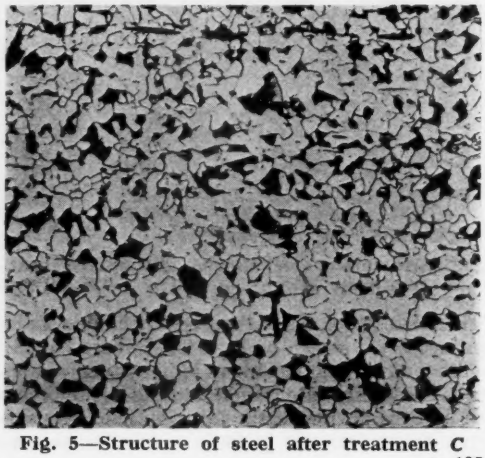


Fig. 5—Structure of steel after treatment *C* × 125

The effect of pearlite structure, as shown by comparison of the results from samples given treatments *A* and *B*, or *G* and *H*, respectively, is that as the pearlite structure becomes finer the transition temperature becomes lower. The effect shown here is not so great as that shown by Grossman,[4] but it must be noted that the techniques used in the heat-treatment of samples differ in important respects; Grossman quenched samples directly from the normalizing temperature into the bath for isothermal transformation, whilst the samples described here were slowly cooled to 710° C to allow some separation of ferrite before being transferred to the bath for isothermal transformation. Early attempts in this work to use direct quenching from the normalizing temperature produced the desired differences in pearlite structure, but also produced substantial differences in ferrite grain size and ferrite–pearlite distribution, which would render a comparison for the effect of pearlite structure open to considerable doubt; it was for this reason that the slow cooling to 710° C, to establish the ferrite grain size and the ferrite–pearlite distribution, was introduced.

Fig. 6—Structure of steel after (*a*) and (*b*) treatment *G*, (*c*) treatment *G'* (*a*) and (*c*) × 125, (*b*) × 750

Cooling rate below the temperature of completion of pearlite formation does not appear to have any great effect on the transition temperature, at least within the range of cooling rates used in these experiments. Rates of cooling which would produce substantial quench-ageing effects were avoided since it was to be expected that quench-ageing would result in a marked rise in transition temperature.

It has already been remarked that comparison of samples given treatments *A* and *G*, and treatments *B* and *H*, is not satisfactory as an indication of the effect of austenite grain size alone, since it has not been possible to produce substantially identical final structures in samples having different initial austenite grain sizes, but it may be worth noting in Fig. 9 that all the samples starting from a coarse austenite grain size (treatments *G*, *H*, and *J*) have much higher transition temperatures than the samples starting from a fine austenite grain size (treatments *A* to *F*). This suggests that the austenite grain size of the material at the normalizing temperature, or at the end of hot working, is a major factor of thermal history affecting the transition temperature; if the austenite grain size at this stage is coarse it seems likely that the transition temperature will be high, whatever the cooling conditions.

There is no indication from this work that any one stage of the cooling range has a predominant effect on the transition temperature, but some inferences as to the probable effect of cooling rate in different parts of the range can be drawn.

In agreement with other work, there is an indication that a fine ferrite grain size is to be desired; this is largely ensured by a fine austenite grain size, but in addition the cooling rate over the temperature range in which the ferrite is separating should be fairly fast. Comparing the ferrite grain sizes shown in Figs. 2*e* and 2*b*, it seems unlikely that from the point of view of ferrite grain size there will be much advantage in exceeding the rate of cooling in still air of a $\frac{1}{2}$-in. plate.

The steel should also be cooled sufficiently rapidly to produce a fine pearlite structure and, for this purpose, it is probable that a cooling rate rather faster than the free rate in still air will be of advantage.

There is no indication that the rate of cooling below the temperature at which pearlite transformation is complete has any significant effect on the transition temperature, provided that cooling in this range is not so fast as to cause quench-ageing effects.

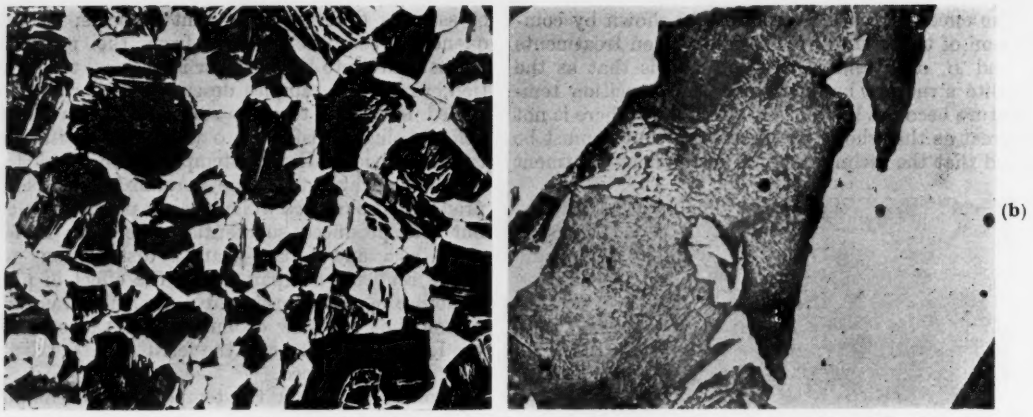


Fig. 7—Structure of steel after treatment *H* (*a*) × 125, (*b*) × 750

SAMPLES NON-UNIFORMLY COOLED

To confirm the validity of the inferences about the effects of cooling rate, a few samples were cooled from 860° C under conditions in which the cooling rate from one part of the range to another was varied. In these experiments the whole cooling range was regarded as composed of three regions, 860–700° C, in which the ferrite grain size and the distribution of the pearlite areas would be established, 700–600° C, in which transformation of austenite to pearlite should be completed at the cooling rates used, and from 600° C down to room temperature.

The cooling rates used for these samples were the same as those used in the uniform cooling-rate experiments described earlier, and the results from those experiments can be used in conjunction with the results from the non-uniform cooling tests to assess the effects of a change of cooling rate at some stage in the range.

The cooling schedules available for comparison are:

Cooling Rate in the Range			
860–700° C	700–600° C	600° C–R.T.	Code
Furnace	Furnace	Furnace	*FFF*
Still air	Still air	Still air	*SSS*
Air blast	Air blast	Air blast	*BBB*
Furnace	Furnace	Air blast	*FFB*
Still air	Air blast	Air blast	*SBB*
Air blast	Furnace	Furnace	*BFF*

The microstructures corresponding to the first three cooling schedules are shown in Figs. 2*d–f*, whilst those for the last three are shown in Figs. 10*a–c*; the transition temperature ranges for the six samples are shown diagrammatically in Fig. 11.

Considering first the microstructures, in samples cooled at the same rate through the range 860–700° C the ferrite grain sizes are closely similar, irrespective of the subsequent cooling rates, and the distributions of the pearlite areas are also similar. The cooling rate over the range 700–600° C makes some difference to the apparent amount of pearlite present; examination at higher magnifications than were used for the photomicrographs shown indicated that cooling rate in this range had a controlling influence on the fineness of structure of the pearlite. The cooling rate below 600° C, in the one case where a comparison can be made (Figs. 2*d* and 10*a*), is without any evident effect on the structure.

For the effect of cooling rate down to 700° C on the transition, comparison can be made of samples *BBB* with *SBB* and samples *FFF* with *BFF*. Comparison of the first pair indicates that a change from still-air cooling to air-blast cooling in this range is without effect on the transition temperature; from the second pair it appears that a change from furnace cooling to air-blast cooling leads to a lowering of the transition temperature range.

The effect of changing from furnace cooling to air-blast cooling over the range from 600° C to room temperature, as shown by comparison of samples *FFF* and *FFB*, is to leave the transition temperature practically unchanged.

The samples treated do not allow direct examination of the effect of cooling rate over the range 700–600° C, but, in view of the lack of effect of cooling rate below 600° C, it seems reasonable to assume that any effect of cooling rate shown in the temperature range from 700° C down to room temperature may be considered as an effect of cooling rate in the 700–600° C range.

The effect of cooling rate over the range from 700° C down to room temperature may be examined by comparing samples *BBB* with *BFF* and sample *SSS* with *SBB*. In both cases the faster cooling rate below 700° C results in considerable lowering of the transi-

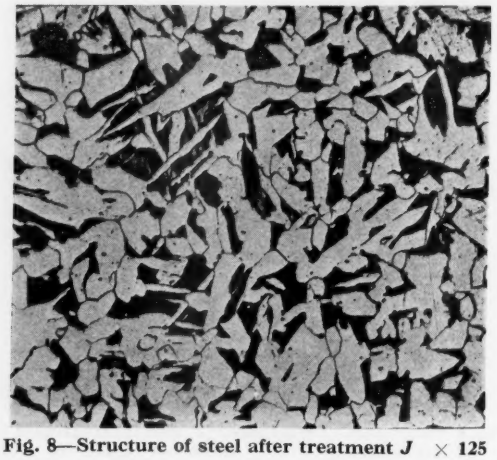


Fig. 8—Structure of steel after treatment *J* × 125

tion temperature and, on the argument put forward above, this may be regarded as mainly, if not entirely, an effect of cooling rate in the 700–600° C range.

SUMMARY AND CONCLUSIONS

Using a ½-in. thick plate of steel containing 0·24% of carbon and 0·55% of manganese, and assessing the transition temperature by the first appearance of cleavage fracture in a notched slow bend test, the main factor affecting the transition temperature of this steel is the austenite grain size at the beginning of the final cooling; if it is coarse the steel will have a high transition temperature, if it is fine a low transition temperature can be attained by suitable cooling.

To cause the steel to show the lowest transition temperature which can be developed from the original austenite grain size the cooling rate down to about 700° C should not be slower than the rate of cooling of a ½-in. thick plate in still air (about 70 °C/min), but there appears to be no advantage in exceeding this rate; the cooling rate from 700° to 600° C can, with advantage, be faster than the rate of a ½-in. plate in still air, and can usefully be as high as 200 °C/min. Below 600° C the rate of cooling appears to have no significance provided it is not so fast as to lead to quench-ageing effects in the steel.

From this work it seems that to obtain a low transition temperature for a given steel it should either finish rolling, or be normalized, at a temperature at which the austenite grain size is fine and that cooling down to 600° C should be fairly fast, perhaps of the order of 200 °C/min, if this is attainable; below 600° C the cooling rate is less important—it can be quite slow and should not be so fast as to give rise to quench-ageing effects.

APPENDIX

Observations on Some Fractures in Notched Slow Bend Tests

During the testing of some specimens it was noted that, while the test piece was still behaving essentially elastically, audible clicks were emitted at intervals, and that each click was accompanied by a small jog

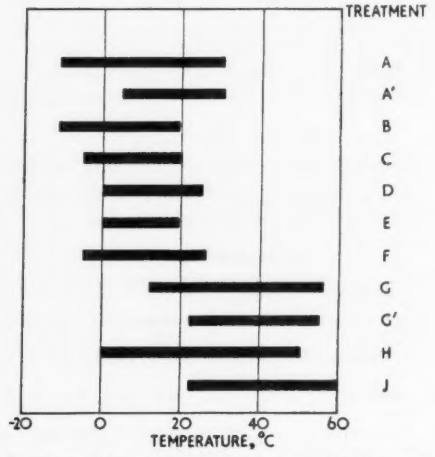
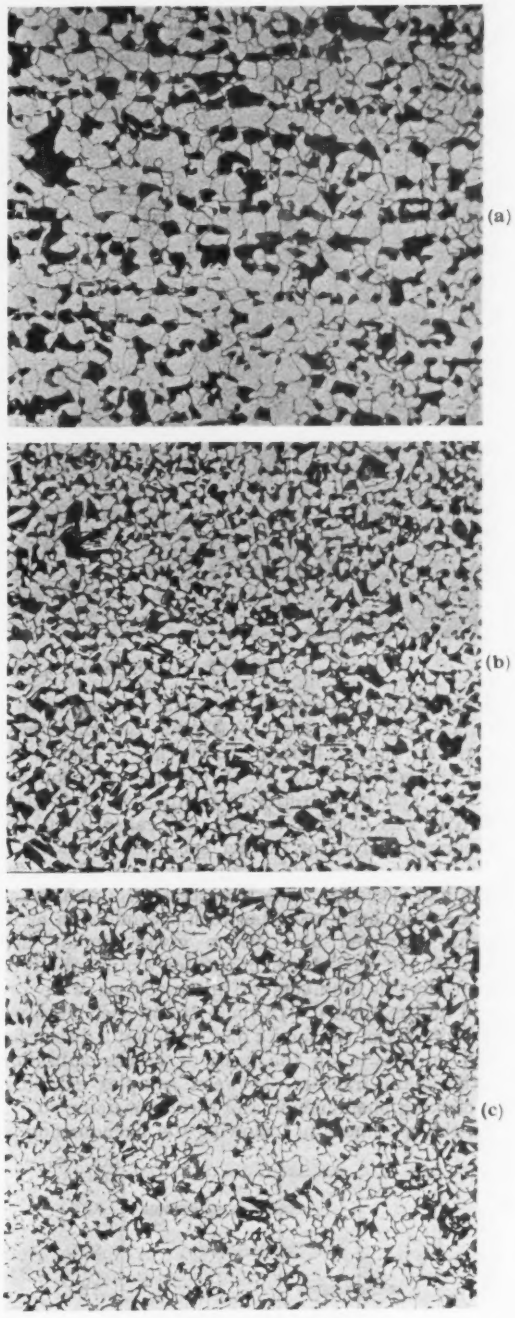


Fig. 9—Transition temperature ranges for samples given special heat treatments



(*a*) Furnace followed by air blast (*FFB*)
(*b*) Still air followed by air blast (*SBB*)
(*c*) Air blast followed by furnace (*BFF*)

Fig. 10—Structure of steels non-uniformly cooled from 860° C × 125

in the autographic stress/strain curve; after this small displacement the curve resumed a straight line.

It was thought that this phenomenon might be due either to twinning of some grains, or to the formation

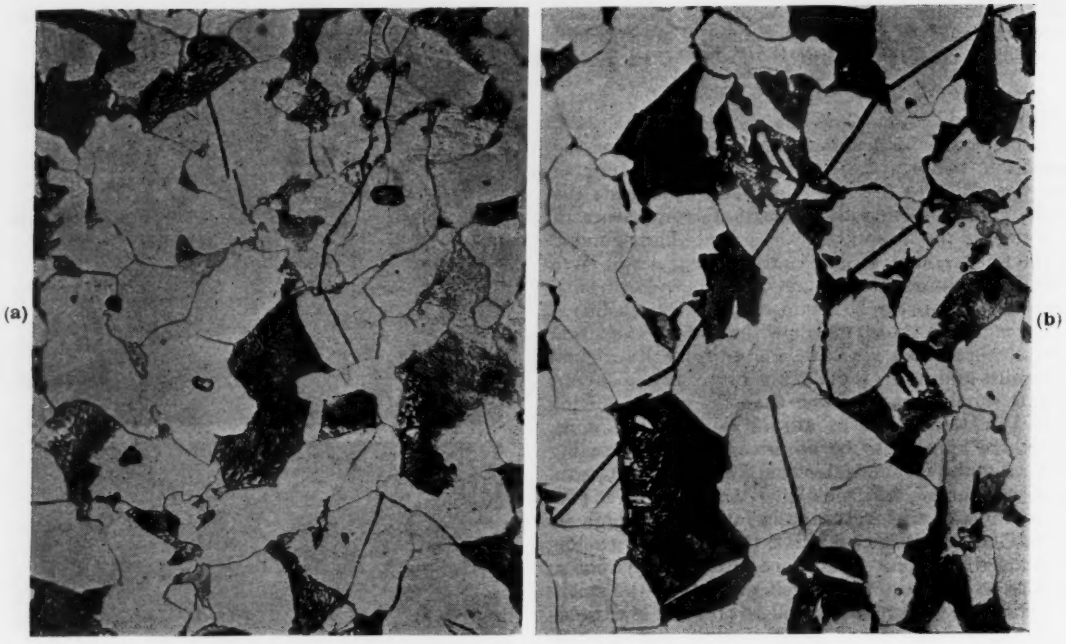


Fig. 12—Cleavage cracks remote from main fracture path × 525

of isolated cleavage cracks in some grains much in advance of general fracture; Tipper and Hall[6] have referred to audible twinning occurring in their tests.

All the specimens in which this phenomenon was observed had been tested to completion of fracture before it was decided to seek the cause of it, and subsequent search for further specimens showing it, in which the test could be discontinued after a few clicks had occurred, was unsuccessful. Comparison has therefore been made between specimens in which clicking occurred and others given similar heat treatments in which the phenomenon was not observed.

Considerable numbers of microsections were prepared and examined. In none of the specimens was any twinning of the ferrite observed, but a distinction could be drawn, on the basis of cracks subsidiary to the main fracture, between samples exhibiting the clicking phenomenon and those not doing so.

In the specimens not showing the phenomenon, all subsidiary cracks found were in the immediate vicinity of the main fracture and appeared to be associated with it; in all specimens in which clicking occurred there were short cracks, sometimes confined to a single grain, sometimes continuing across a few grains, at some distance from the main fracture path and appearing not to have any connection with it; examples of cracked single grains and of cracks traversing a few adjacent grains are shown in Figs. 12*a* and 12*b*. The cracks appear to be cleavage type.

From this examination it appears that the audible clicking occurring during the straining of some specimens, and which was accompanied by small irregularities in the stress/strain curves, was not due to twinning of the ferrite grains but probably to cleavage cracking; this occurred at low values of general strain in isolated grains, the cracks being arrested after traversing a very few grains at the most, and sometimes being confined to a single grain.

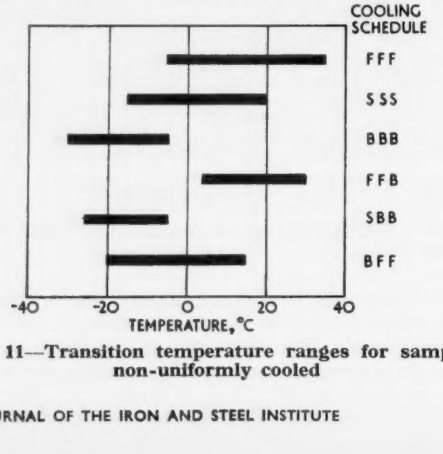


Fig. 11—Transition temperature ranges for samples non-uniformly cooled

Acknowledgment

The work described in this paper was carried out at Admiralty Materials Laboratory, and is published by permission of the Admiralty.

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The Absorptiometric Determination of Copper in Iron and Steel

The B.I.S.R.A. Methods of Analysis Committee

SYNOPSIS

The diethyldithiocarbamate and diquinolyl absorptiometric methods for copper have been critically examined and the latter is recommended as a standard procedure for all types of iron and steel.

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TOWARDS THE END of 1952 attention was drawn to the absence of an accepted referee method for the absorptiometric determination of copper in iron and steel and the B.I.S.R.A. Physico-Chemical Methods of Analysis Sub-Committee* was asked to conduct an investigation on this subject.

Among recently published colorimetric procedures, the following references were noted: (i) the direct diethyldithiocarbamate method,[1] (ii) the extraction diethyldithiocarbamate method,[2] and (iii) the 2–2′ diquinolyl method.[3] A programme in three sections was therefore adopted.

DIETHYLDITHIOCARBAMATE PROCEDURES

Direct Measurement

The published method[1] involved solution of the sample in sulphuric–phosphoric acid mixture, oxidation with nitric acid, and coloration of an aliquot portion with sodium diethyldithiocarbamate in ammoniacal solution in the presence of gum acacia. Metals insoluble in aqueous ammonia were held in solution by means of ammonium citrate, and nickel interference was suppressed by the addition of excess ammonium hydroxide. The resultant colour was measured on an absorptiometer with blue filters and a tungsten lamp.

The range of the first series of results obtained by six analysts was:

Sample	Copper, %
MGS/21 (mild steel)	0·105–0·115
MGS/38 (0·25% Cu mild steel)	0·24 –0·26
MGS/75 (0·5% Cu pig iron) ...	0·50 –0·53
MGS/89 (0·5% W steel) ...	0·11 –0·12

These and other tests, while confirming the satisfactory operation of the method, brought to light a few minor difficulties, including the wide variation in quality of gum acacia and the interference of cobalt.

However, the situation was sufficiently encouraging to warrant further investigation of the effect of alloying elements and several synthetic solutions were examined. It was appreciated that other mineral acids might be required for the decomposition of alloy steels, being subsequently removed by fuming, and this step was introduced for this series of tests.

Only cobalt and high manganese contents were found to influence the copper determination, and it was at first thought that the interference of both elements could be overcome by reduction with sulphurous acid before colour development, since extraneous colour appears to be produced by traces of cobalt or manganese compounds which were not in the fully reduced state. Further investigation indicated that the trouble was not completely overcome and that slight turbidities could occasionally result from the presence of these elements. The use of the reagent for the determination of cobalt and nickel as well as copper had already been reported.[4]

It was felt that the direct method, although applicable to a considerable range of steels and adequate for routine control purposes, was not of sufficiently general application for a referee method. An extraction method was therefore considered.

Organic Solvent Extraction

The main differences between this and the direct method are the use of disodium ethylenediamine tetra-acetate in a slightly ammoniacal solution to overcome interferences, and an organic solvent for extracting the coloured complex. In the original procedure,[2] butyl-acetate extraction was recommended, but since this solvent is lighter than water, carbon tetrachloride was tried in the preliminary work.

The range of results obtained in preliminary trials was:

Sample	Copper, %
MGS/151 (14% Cr, 14% Ni, 10% Co, 3% Nb, 2·5% W, 1·75% Mo, 1·5% Si)	0·07–0·09
MGS/166 (4% V high-speed steel) ...	0·17–0·20
MGS/167 (4% Mo high-speed steel) ...	0·11–0·145
MGS/237 (1% Nb permanent magnet alloy)	3·0 –3·3
MGS/243 (18/8/1% Cu stainless steel)	0·95–1·03
MGS/244 (0·5% Cu mild steel)	0·52–0·56

* The composition of the Physico-Chemical Methods of Analysis Sub-Committee, Metallurgy (General) Division, B.I.S.R.A., on 30th June, 1954, was:

Mr. H. G. Short (*Chairman*), National Physical Laboratory; Mr. H. Cox, Metropolitan-Vickers Electrical Co. Ltd.; Mr. E. Dyke, Brown-Firth Research Laboratories; Mr. F. E. Eborall, G. Senior and Sons Ltd.; Mr. L. E. Gardner, Elgar Allen and Co. Ltd.; Mr. C. H. R. Gentry, Philips Electrical Ltd.; Mrs. L. E. Greenfield, G.K.N. Group Research Laboratories; Mr. H. Groom, Bragg Laboratory, N.O.I.D.; Mr. S. Harrison, Kayser Ellison and Co. Ltd.; Dr. T. S. Harrison, Appleby-Frodingham Steel Co.; Mr. G. Padget, United Steel Companies Ltd.; Mr. D. F. Sermin, S. Osborn and Co Ltd.; Mr. J. O. Lay (*Secretary*), B.I.S.R.A.

The Committee wishes to express its appreciation of the work of Mr. D. Bell (Colvilles, Ltd.) and Miss R. Presser (British Cast Iron Research Association), during the early stages of the investigation.

Table I

NORMAL COMPOSITION OF SAMPLES, %

Sample	C	Mn	Ni	Cr	Mo	V	Co	W	Nb	Others
MGS/21	0·3	1·6	0·2	0·2	0·05	...	...	...	...	...
MGS/38	0·3	0·8	0·1	0·1	0·03	...	...	...	...	...
MGS/49	0·8	0·4	...	4	...	1	...	18	...	0·14 Sn
MGS/75	3·3	0·7	...	...	0·3	...	...	...	...	...
MGS/89	1·0	0·8	...	0·8	0·02	...	...	0·5	...	...
MGS/151	...	0·8	13·4	13·5	1·65	...	10	2·5	3	...
MGS/166	...	...	...	4	...	4	...	10	...	...
MGS/167	...	...	...	4·5	4	1·5	...	6	...	...
MGS/215	3·1	1·0	...	...	...	...	...	...	...	...
MGS/237	...	...	18·5	...	...	...	12	...	...	9 Al
MGS/243	0·1	0·5	8·7	19·0	0·04	0·02	...	0·7	...	0·6 Ti
MGS/244	0·2	1·6	0·1	0·1	0·01	...	...	...	...	...

Owing to difficulties in obtaining complete extraction of the copper complex some modifications to the experimental conditions were made, but several extractions with solvent were still required. Some improvement was achieved with butyl acetate in place of carbon tetrachloride, but this requires repeated extractions with a solvent lighter than water.

In addition, the main disadvantage of the direct method, viz. the interference of cobalt and high manganese, was not completely eliminated, although much reduced, and attention was transferred to the use of 2–2′ diquinolyl, which was reported to be specific for copper.

DIQUINOLYL PROCEDURE

The simple technique adopted for the use of this reagent[3] was a major attraction. In an oxidized solution of the sample in phosphoric–sulphuric acid, copper is reduced to the cuprous state with hydroxylamine, the solution is buffered with ammoniacal citrate solution, and the copper complex is formed by adding the diquinolyl reagent dissolved in amyl alcohol. The alcohol is then separated, carrying the copper complex in solution, and the colorimetric measurement made.

This procedure was tested on the first four samples circulated and reproducibility at least as good as that with diethyldithiocarbamate was found. Some slight difficulties were experienced due to a supply of inferior reagent and to the re-oxidation of the sample solution unless the reagent was added and readings taken within a fairly short time after addition of the hydroxylamine. There was also some indication that tungsten would interfere, but an increase in the amount of phosphoric acid in the solvent prevented error from this source.

Comprehensive tests were made with the method finally accepted (*see* Appendix) on a series of samples, the composition of which is indicated in Table I. The results in Table II are very satisfactory and show good agreement with previous chemical determinations where figures are available. The relative reliability of the diethyldithiocarbamate method, using extraction where necessary, is also indicated.

The diquinolyl method has been deliberately limited to an upper copper content of 1% (although a sample has been analysed containing 3% copper) since this is felt to be the upper limit for a colorimetric method using the conditions described. Volumetric or other methods are recommended for referee analysis of steels containing more than 1% copper.

CONCLUSIONS

The direct diethyldithiocarbamate colorimetric method for copper in ferrous material is suitable for low-alloy compositions but interference is caused by the presence of cobalt and high manganese. By solvent extraction of the coloured copper complex in the presence of ethylenediamine tetra-acetate these interferences can be considerably reduced, but experimental conditions must be carefully chosen to obtain complete extraction and reproducible results.

The 2–2′ diquinolyl reagent appears specific for copper, and although requiring an extraction procedure offers considerable advantage over the former method in ease and reliability of operation. A method employing this reagent is described in the Appendix and will be recommended for consideration as a British Standard.

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APPENDIX

Recommended Method

Principle—Copper in a phosphoric–sulphuric solution of the sample is complexed with 2–2′ diquinolyl and the resultant coloured compound extracted in amyl alcohol. The copper content is determined by measurement of the absorption and evaluation from a calibration graph.

Range—Up to 1% of copper (Note 1).

Reproducibility—Average results at the 0·1% copper level should not differ by more than ± 0·01% of copper between laboratories, and at 1% by not more than ± 0·03%.

This method has been found satisfactory in the presence of cobalt, manganese, nickel, tungsten, and other elements.

SPECIAL APPARATUS

A satisfactory instrument for this method is the Hilger photo-electric absorptiometer. Using this instrument, suitable conditions for the determination of copper are:

Mercury vapour lamp
Wratten 74 with Ilford H503 filters
1-cm cells.

Other instruments suitable for measuring the optical density of the solution at 540–560 mμ may be employed, but slight changes in operating technique may be required. Grade A graduated glassware shall be used throughout.

Table II

ANALYSIS BY RECOMMENDED METHOD (COPPER, %)

Analyst	MGS/21	38	49	75	89	151	166	167	215	237	243	244
1	0·112	0·24	0·55	0·505	0·107	0·085	0·17	0·127	0·015	3·30	0·99	0·54
2	0·122	0·25	...	0·51	0·12	...	...	...	...	...	...	...
3	0·117	0·245	0·59	0·53	0·120	0·092	0·18	0·132	0·015	3·29	1·01	0·54
4	0·114	0·24	0·56	0·51	0·112	0·087	0·17	0·126	0·015	3·09	0·99	0·53
5	0·115	0·24	0·57	0·51	0·113	0·091	0·18	0·133	0·019	3·23	1·00	0·54
6	0·116	0·25	0·58	0·52	0·114	0·086	0·18	0·134	0·015	3·18	1·02	0·55
7	0·115	0·25	0·58	0·53	0·114	0·088	0·19	0·135	0·020	3·31	1·03	0·56
8	0·118	0·255	0·585	0·52	0·117	0·095	0·18	0·128	0·021	3·18	0·99	0·54
9	0·116	0·255	0·585	0·53	0·118	0·092	0·195	0·125	0·017	3·36	1·00	0·57
10	0·113	0·25	0·565	0·515	0·114	0·085	0·185	0·120	0·016	3·09	0·98	0·54
11	0·116	0·25	0·58	0·52	0·117	0·091	0·18	0·129	0·016	3·24	1·01	0·56
Average	0·116	0·25	0·57	0·52	0·115	0·089	0·18	0·129	0·017	3·23	1·00	0·55
Accepted Figure	...	0·245	...	...	...	0·09	0·18	0·13	...	...	0·98	...
Diethyldithiocarbamate Average	0·112	0·25	...	0·51	0·114	0·084	0·185	0·132	...	3·12	0·97	0·54

REAGENTS REQUIRED

All reagents shall be of the highest purity obtainable and distilled water shall be used throughout. Solutions shall be freshly prepared and filtered where necessary.

Solutions

Phosphoric–Sulphuric Acid (15/15%v/v)

Carefully pour 150 ml of phosphoric acid (sp. gr. 1·75) followed by 150 ml of sulphuric acid (sp. gr. 1·84) into about 600 ml of water, cool and dilute to 1 l., stirring thoroughly meanwhile.

Ammonium Citrate

Dissolve 500 g of citric acid in about 200 ml of water, cautiously add 500 ml of ammonium hydroxide (sp. gr. 0·880), cool and dilute to 1 l.

Diquinolyl

Dissolve 0·05 g of 2–2′ diquinolyl in 100 ml of amyl alcohol.

Sampling

Recommended methods of obtaining a suitable sample for the analytical procedure outlined below are described in B.S. 1837.

Procedure

Transfer 0·5 g of sample (Note 1) to a 250-ml beaker, dissolve in 10 ml of phosphoric–sulphuric acid (15/15% v/v) (Note 2) by heating and oxidize with 2 ml of nitric acid (Note 3). Evaporate to fumes, cool, dilute to about 30 ml. Warm if necessary, to dissolve all soluble salts, and cool. Transfer to a graduated flask and dilute to 50 ml, mix well.

Pipette 10 ml into a 100-ml beaker, add 1 g of hydroxylamine hydrochloride and warm on a steam bath for 5 min. Cool quickly to below 20° C, add 10 ml of ammoniacal citrate solution, and wash into a 50-ml separating funnel with a minimum of water. Add immediately (Note 4) from an automatic pipette or fast-running burette 10 ml of diquinolyl solution, shake for 1 min, allow to separate and discard the aqueous layer. Fill a 1-cm cell by running the alcohol layer through a Whatman 41 filter paper (Note 5) and measure the absorption against a water cell (Note 6).

Calibration

Copper solution A—Dissolve 2·0 g of pure copper in 30 ml of nitric acid (40% v/v), boil to remove nitrous fumes, and dilute to 1 l. (1 ml = 0·002 g of copper).

Copper solution B—Dilute 25 ml of copper solution *A* to 1 l. (1 ml = 0·00005 g of copper).

Procedure

To several 0·5 g samples of pure iron (Note 7) in 250-ml beakers, add volumes of copper solution *B* to give an appropriate range (1 ml = 0·01% Cu on a 0·5 g sample) and treat as described in the " Procedure " section above. Prepare a calibration graph by plotting the absorptiometer readings against copper content of the standards.

The copper content of test samples is then read direct from the calibration graph.

NOTES

(1) For copper contents over 0·1% adjust the sample weight as follows:

0·1–0·5% Cu—0·1 g sample

0·5–1% Cu—0·1 g sample with 30 ml of phosphoric–sulphuric acid (15/15% v/v), diluting to 200 ml after fuming.

(2) For samples containing tungsten, it is necessary to add 1 ml of phosphoric acid (sp. gr. 1·75) extra to the solvent acid for every 0·01 g of tungsten present.

(3) Hydrochloric acid may be employed where necessary to ensure decomposition of the sample, but should be completely removed by fuming.

(4) The sequence of operations following reduction with hydroxylamine should be carried out as quickly as possible until the alcohol extraction is complete in order to avoid re-oxidation and contamination of the extract.

(5) If evaporation of the alcohol is prevented, the coloured solution remains stable for several hours.

(6) It is essential to carry out a blank determination on the reagents, using copper-free pure iron, but a compensating solution is unnecessary.

(7) If copper-free pure iron is not obtainable, an appropriate zero correction must be made.

British Iron and Steel Research Association

COMMITTEE ORGANIZATION CHART *at December 31st, 1955*

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  - AA Sintering
- B Process
- C Practice
- D Blast-furnace refractories (*joint with B.Ceram.R.A.*)
- E Coke (*joint with B.Coke R.A.*)

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  - DF Alumino-silicate refractories
  - DG Silica refractories

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  - AA Roll design conference
- B Solid mechanics group
- C Coatings
  - CA Spent pickle liquor
- D Wire drawing
- E Sheet metal working
  - EA Cupping press
- F Forging

PE PLANT ENGINEERING

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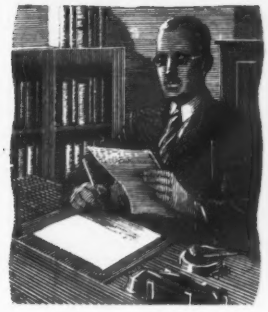
- A Furnaces and fuel
- R Rolling-mill plant
- H Mechanical handling plant
- N New engineering techniques

MG METALLURGY GENERAL

- K Structural engineering steels
- H Electrical sheet
  - HA Electrical sheet analysis
- L Tool steels
- J High-temperature steels (*joint with E.R.A.*)
  - JA Scaling of steels at high temperatures
- A Alloy steels
- M Fatigue and inclusions
- N Thermal treatment
  - NA Isothermal transformations and hardenability
- D Methods of analysis
  - DA Raw materials and slags
  - DB Iron and low-alloy steels
  - DC High-alloy steels
  - DD Spectrographic
  - DE Physico-chemical
- O Castaing micro-analysis technique (*joint with B.C.I.R.A., B.N.F.M.R.A., and B.W.R.A.*)
- B Corrosion
  - BA Protective coatings (corrosion)
  - BB Atmospheric corrosion
  - BD Waters (corrosion)
  - BE Joint committee on the corrosion of buried metals
  - BF Methods of testing (corrosion)

CORRESPONDENCE

Correspondence on Papers · Letters to the Editor



Correspondence on the Paper—

THE CORROSION RESISTANCE OF WROUGHT IRON*

By J. P. CHILTON and U. R. EVANS

Dr. **K. Sachs** (G.K.N. Research Laboratory) wrote: The authors have provided a definite explanation of the corrosion behaviour of wrought iron and have shown that the traditional belief that wrought iron is more resistant to corrosion than steel has some basis in fact, even if it is difficult to corroborate by laboratory experiments. The presence of alloy segregates (*V* zones) accounts for localized high corrosion resistance, and the authors suggest that some of the advantages of wrought iron might be secured by the use of piled steel, which would be much more convenient to produce. Without slag in the structure the presence of *Q* and *R* zones cannot be expected and the corrosion resistance of the material could only depend on the *V* zones. I am inclined to question the probability of the formation of *V* zones by the oxidation of mild steel, in the absence of a liquid phase.

The authors' explanation of the origin of *V* zones in wrought iron, on p. 118 of their paper, implies that the oxides on the surface have to be molten. Some years ago I did some experiments on the scaling of nickel steel in the laboratories of the Mond Nickel Co., Ltd., which serve to confirm that layers enriched in alloy content can be formed on the surface of steel, and also underline the need for the presence of a liquid phase.

The structure of the scale formed on a 1½% Ni steel in 24 h at 1250° C is shown in Fig. *A*. Metal particles are entangled with the oxide and it has been shown that these particles are enriched in nickel. On one unhappy occasion such a specimen was placed on a chrome-magnesite brick and scaled for 24 h at 1250° C. The iron oxides of the scale reacted with the chromic oxide of the brick to form a spinel; the scale was fluxed and on removal from the furnace the remainder of the specimen

J. Iron Steel Inst., 1955, vol. 181, Oct., pp. 113–122.

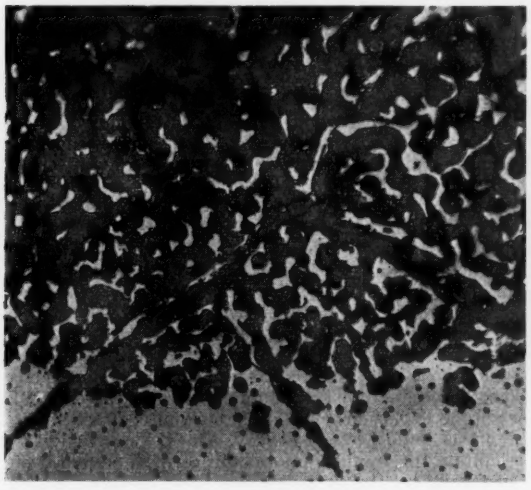


Fig. A—Scale formed on 1½% Ni steel in 24 h at 1250° C

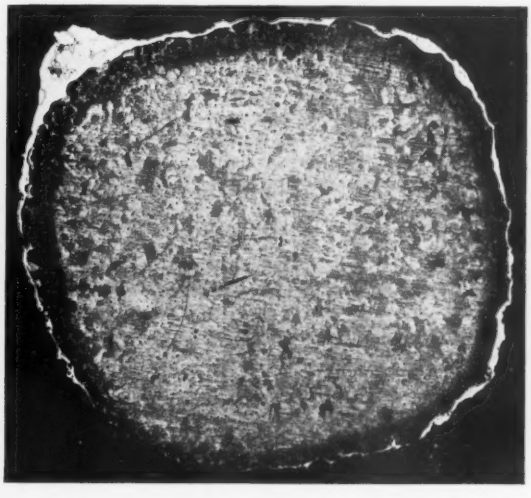


Fig. B—Cross-section of specimen scaled for 24 h at 1250° C on chrome-magnesite brick

was found deeply buried in the brick. When the latter was broken open and the specimen extracted, all that was left was a grey metallic slug, cylindrical with rounded ends, having a thin oxide coating.

The cross-section is shown in Fig. *B*. The specimen was etched electrolytically in 5% sulphuric acid and a thin white surface layer can be clearly distinguished. Spectrographic analysis indicated the presence of about 10% Ni in this layer. The deposit is thicker on one side than the other and there is a large ' drop ' at one point. The distribution of the nickel-rich material is consistent with deposition from a viscous liquid phase.

Surfaces of the type shown in the second illustration might weld together, forming an alloy-rich layer between the piles, i.e. a *V* zone. Surfaces with solid scales, on the other hand, would only weld after the scale has been cracked off; the alloy-rich particles would disappear with the scale and there would be no *V* zone at the weld.

AUTHORS' REPLY

Dr. **J. P. Chilton** and Dr. **U. R. Evans** (Cambridge University) wrote in reply: The information provided by Dr. Sachs is most instructive and will arouse much interest. We agree that the formation of *V* zones by the piling of mild steel would be difficult, in the absence of slag. It should not, however, too easily be assumed that the difficulties cannot be overcome. Swedish iron, which contains no slag, has been successfully piled. Presumably the iron oxide present on the components of the pile was squeezed out as liquid during the piling process. Examination of such material has revealed what appear to be *V* zones, since at each piling plane a double wall of metal is left behind on corrosion by acid containing hydrogen sulphide; the fact that the wall was double, however, does suggest that the welding had been imperfect.

Correspondence on the Paper—

CAUSES OF VARIABLE CREEP STRENGTH IN BASIC O.H. CARBON STEEL*

By W. E. BARDGETT and M. G. GEMMILL

Dr. **M. L. Becker** (B.I.S.R.A.) wrote: This interesting paper illustrates once again the influence of ladle additions of aluminium on the creep resistance of basic O.H. mild steels. The data given in Table II and Fig. 1 show most convincingly that creep rate as measured in the tests is closely linked with the content of soluble aluminium; and the authors have sought to determine by what mechanism this occurs. The curve shown in Fig. 2 certainly suggests that the creep rate of steel 47 is directly related to the quantity of aluminium present as aluminium nitride but I do not consider that this had been proved; I believe that the austenite grain size at the heat-treatment temperature is more likely to be the controlling factor. The results given in Table VI, where the four steels in question have been tested after heat-treatments producing approximately the same austenite grain size, seem to confirm this view. Quite possibly the coarsening of the austenite could be attributed to the dissociation of the aluminium nitride, but the grain size and not the content of aluminium nitride could be the controlling factor. For example, steel 46, coarsened at 1060° C and air-cooled, gave a low creep rate (8·1) as compared with a high rate (113·8) for the same steel heated to 920° C and air-cooled. Had the material coarsened at 1060° C been cooled and held at 920° C for a short time before air-cooling, the ferrite/pearlite structure would probably have been altered but the aluminium nitride content would have been raised to the original value appropriate to 920° C. Results obtained by other investigators of this type of material, however, indicate that in spite of these structural changes, the creep rate would have remained low despite the increase in aluminium nitride. The latter could not therefore be the primary factor controlling creep rate.

The McQuaid–Ehn grain size given in Table IV of steel 47, heated at increasing temperatures, does not throw much light on this subject but if the actual grain size of the austenite for each temperature could be plotted against creep rate it seems probable that a relationship would be obtained equally satisfying to that shown in Fig. 2.

AUTHORS' REPLY

Mr. **W. E. Bardgett** and Mr. **M. G. Gemmill** (United Steel Companies Ltd.) wrote: Dr. Becker states that the austenitic grain size is probably the major factor affecting the creep resistance, in contrast to the authors' view that this property reflects the influence of aluminium nitride. The results of duplicate tests on steel 32, given in Table V of the paper, appear to support the latter opinion, since the widely differing creep resistance of the two specimens relates more directly to their respective AlN contents, than to their similar McQuaid–Ehn grain sizes.

The effect of austenitic grain size on the creep strength can be considered independently of the formation of AlN in two ways. Tests could be made using specimens cooled directly from solution temperatures above that for complete solution of AlN in the austenite; alternatively specimens could be held at, say, 920° C long enough to precipitate the equilibrium amount of AlN, having been transferred to this temperature from several higher solution temperatures. This latter approach is now being examined.

* *J. Iron Steel Inst.*, 1955, vol. 179, Mar., pp. 211–219.

Letter to the Editor

The Isolation of Silicon Nitride from a Commercial 4% Silicon Steel

THE NITROGEN STUDY GROUP of the British Iron and Steel Research Association, which over the past few years has been re-examining the whole problem of the chemical determination of nitrogen in steel, has recently isolated silicon nitride from a commercial grade of 4% Si steel.

Two samples of nominally 4% Si steel were analysed for nitrogen in the form of bar and the results on both samples, using chemical fuming methods, were in good agreement amongst themselves and with those by the vacuum-fusion method at about 0·005% and 0·013% respectively. The proportion of acid-resistant nitrogen was quite small, 0·001%, in both samples.

Another sample of 4% Si steel was then analysed in sheet form and the nitrogen figures determined chemically ranged between 0·001% and 0·002%, the residue separated by centrifuge failing to yield any significant amount of nitrogen on fuming. It was concluded from this work and the earlier results on the other samples of silicon steel that full nitrogen recovery had been obtained. However, when this sheet sample was analysed by the vacuum-fusion method at the National Physical Laboratory it was shown that the true nitrogen content of the sample was 0·0055%, a result which was subsequently confirmed in other laboratories operating this method.

The sample which gave these anomalous results differed in two respects from the others. Firstly, it was in the form of sheet, not bar, and secondly, no aluminium had been added during manufacture. Other casts, with and without aluminium additions, were therefore analysed at various stages of production from ingot to sheet. With aluminium present the results were consistent throughout, but in its absence the chemical methods again indicated an apparent drop in the nitrogen content of the material in sheet form. The vacuum-fusion method showed, however, that the nitrogen content remained the same as in the earlier production stages. In the absence of aluminium, therefore, a large proportion of the nitrogen in this type of steel is converted into an acid-insoluble form during the last reheating or hot-rolling operation.

The residue from the sheet specimens, obtained during chemical analysis, was not amenable to X-ray analysis owing to precipitated silica, but ultimately, by modifications to the solution procedure to avoid this precipitation, a clear X-ray photograph of the residue was obtained. The pattern was identical with that of synthetically prepared silicon nitride (Si_3N_4) and it is believed that this is the first time that this nitride has been positively identified in a commercial steel.

Silicon nitride has been found to be remarkably resistant to attack by acids and the Nitrogen Study Group is now attempting to find ways of obtaining full nitrogen recovery from steels containing this compound by chemical methods. The most hopeful line of attack would appear to be to re-convert the nitrogen, by suitable heat-treatment, into the soluble form in which it exists during the earlier stages of production.

H. A. SLOMAN (National Physical Laboratory: Metallurgy Division)

Chairman, Nitrogen Study Group, British Iron and Steel Research Association.

A Wire Temperature Meter

BY J. MANUEL, B.SC.

A WIRE TEMPERATURE METER which measures the temperature of a moving wire has been developed by the Physics Department of the British Iron and Steel Research Association for application in wire-drawing research. The instrument utilizes a resistance thermometer coil, mounted in a suitable housing (Fig. 1) and connected into one arm of a Wheatstone bridge which is adjusted to balance when the coil is at room temperature. On passing through the coil, the hot wire raises its temperature and hence unbalances the bridge. By using a wire at a known temperature the out-of-balance current has been related to the wire temperature. This calibration is independent of the speed at which the wire is moving, speeds of up to 1000 ft/min being used in the calibration. The instrument is accurate to ±7° C.

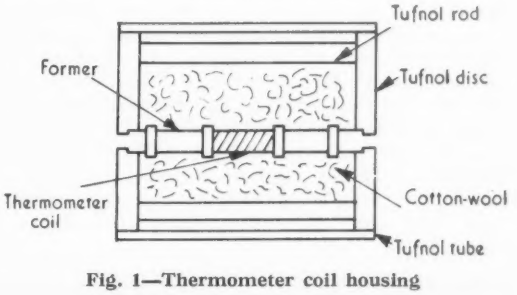
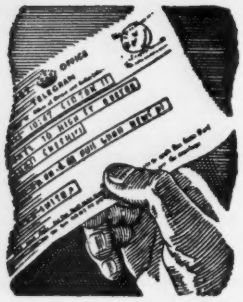


Fig. 1—Thermometer coil housing

Manuscript received 5th November, 1955.

Mr. Manuel is with the Physics Department of B.I.S.R.A.



N E W S

Announcements and News of Science and Industry

IRON AND STEEL INSTITUTE

Annual General Meeting, 1956

The Annual General Meeting of the Institute will be held in London on Wednesday and Thursday, 16th and 17th May, 1956. The dinner for members will be held at Grosvenor House on the evening of 16th May.

Belgian Decoration for the Secretary

His Majesty Baudouin, King of the Belgians, has conferred on the Secretary, Mr. Kenneth Headlam-Morley, the Decoration of *Chevalier de l'Ordre de la Couronne*. The award is in recognition of Mr. Headlam-Morley's services as General Secretary of the European Committee of the Congrès Internationale de Métallurgie, i.e. the Joint Metallurgical Societies Meeting in Europe 1955, in which the Centre National de Recherches Métallurgiques and the Association des Ingènieurs sortis de l'École de Liège participated actively. The Royal Decree is dated 9th December, 1955. The Decoration was handed to Mr. Headlam-Morley at the Belgian Embassy in London by the Marquis du Parc Locmaria, the Belgian Ambassador at the Court of St. James.

NEWS OF MEMBERS

R. J. Brown

The appointment of Mr. R. J. Brown is that of Materials Engineer to the British Motor Corporation, which embraces the Austin Motor Co., Ltd. Appointment to the latter Company was stated in error in the January issue.

C. J. Dadswell

It was stated in error in the January issue that Dr. C. J. Dadswell had recently been appointed a Director of the English Steel Corporation Ltd. Dr. Dadswell has held this appointment since 1946 and the announcement should have referred to his becoming a Director of the Darlington Forge Ltd.

NOTICE TO READERS

From time to time comments on the presentation and contents of the *Journal* reach the Institute, either by letter or verbally. These are always examined with great care by the Editor and the Publication Committee. Improvements resulting from such suggestions have in the past been embodied in the *Journal*. However, although many members and readers must have opinions about how the *Journal* might be improved, these are seldom brought to the notice of those responsible for its preparation. The Council therefore wishes it to be made known that suggestions concerning the presentation and contents of the *Journal* are always welcomed from any reader and will be given the most sympathetic consideration; they should be sent in writing to the Editor at the Offices of the Institute.

Mr. **A. S. Appleton** has been awarded the degree of B.Eng. in Liverpool University.

Mr. **S. Ardmongon** has been awarded the degree of B.Sc. in Durham University and has taken up an appointment with the Thai Naval Attaché, London.

Mr. **F. Atkinson** has retired from the Directorship of Brown Bayley Steels Ltd., Sheffield.

Mr. **F. W. J. Bailey** has been appointed Lecturer in Metallurgy at the North Gloucestershire Technical College, Cheltenham.

Mr. **A. M. Bankier** has been awarded the degree of B.Sc. in Durham University and has taken up a post of Chemist with the Ministry of Supply Royal Ordnance Factory, Birtley, Co. Durham.

Mr. **E. G. Bayliss** has taken up an appointment with Forgings and Presswork Ltd., Birmingham.

Mr. **E. Bishop** has been appointed Assistant to the Director of Research, B.S.A. Group Research Centre, Sheffield.

Dr. **J. C. Blade** has taken up an appointment with the Aluminium Laboratories Ltd., Banbury, Oxfordshire.

Mr. **J. E. Bowers** has been awarded the degree of Ph.D. in Sheffield University.

Mr. **B. T. Bradbury** has been awarded the degree of B.Met. in the University of Sheffield.

Mr. **C. D. S. Bridgett** has been appointed Metallurgist and Industrial Chemist with the R.E.M.E. Maintenance Technique Development Establishment, Chislehurst, Kent.

Mr. **D. A. Brown** has been appointed Chemistry Master at the Stationers' Company's School, London.

Mr. **J. R. A. Bull**, Managing Director of Sanderson Brothers and Newbould Ltd., has been appointed Chairman and Managing Director.

Mr. **G. Bullock** has been awarded the degree of Ph.D., in the University of Manchester and has been appointed Lecturer in Physical Metallurgy at Birmingham University.

Mr. **F. R. Cowlishaw** has been appointed Works Metallurgist with the Dart Spring Co., Ltd., West Bromwich.

Sir **John Craig,** formerly Chairman and Joint Managing Director, has been appointed Honorary President of Colvilles Ltd.

Mr. **D. W. O. Dawson** has been appointed Chief Inspector with A. P. V. Paramount Ltd., Crawley, Sussex.

Mr. **F. Devine** has been appointed Senior Consultant to Associated Industrial Consultants Ltd., London.

Mr. **P. R. Dixon** has been awarded the degree of M.Met. in Sheffield University.

Mr. **R. Doldon** has been awarded the degree of B.Met. in Sheffield University and has been appointed Scientific Officer with the U.K. Atomic Energy Authority, Research and Development Branch, Culceth, Lancs.

Dr. **V. W. Eldred** has been appointed Principal Scientific Officer with the U.K. Atomic Energy Authority, Industrial Group, Research and Development Branch, Sellafield.

Mr. **J. E. Elliott** has taken up a post as Assistant Metallurgist with Richard Johnson and Nephew Ltd., Manchester.

Mr. **R. P. P. Ellis** has taken up an appointment as Metallurgist with George Wimpey and Co., Ltd., Southall, Middlesex.

Mr. **F. C. Evans** has been appointed Managing Director of Metallurgical Engineers, London.

Mr. **G. J. Evans** has taken up a post as Metallurgist with Enfield Rolling Mills (Aluminium) Ltd., Enfield, Middlesex.

Mr. **J. R. Evans** has taken up the post of Junior Metallurgical Observer with the Steel Company of Wales, Ltd., Port Talbot.

Dr. **U. R. Evans,** F.R.S., has been awarded the Palladium Medal of the Electrochemical Society.

Dr. **P. M. Fisk** has been appointed a Lecturer at the Borough Polytechnic, London.

Mr. **M. R. Fox** has taken up a post as Metallurgical Assistant with Bayliss, Jones and Bayliss Ltd., Wolverhampton.

Mr. **A. M. Frankau** has taken up an appointment as Technical Assistant with Huntington, Heberlein and Co., Ltd., London.

Mr. **K. Guttenstein** has been appointed Chairman and Joint Managing Director of the Loewy Engineering Co., Ltd., Bournemouth.

Mr. **C. C. Hanson** has been appointed Chief Metallurgist of Radiation Ltd.

Mr. **R. J. Harbord** has been awarded the degrees of B.Sc. and A.R.S.M. at London University and has taken up an appointment with Riley, Harbord and Law, London.

Mr. **F. Hargreaves** has taken up an appointment with the Chemical and Metallurgical Division of the Plessey Co., Ltd.

Mr. **G. C. Harper** has been appointed a Director of C. S. Milne and Co., Ltd., London.

Mr. **J. D. Harper** has been awarded the degree of B.A. (Cantab.).

Mr. **L. Henderson** has been awarded the degree of B.Sc. in the University of Durham and has been appointed Metallurgist with Clarke, Chapman and Co., Ltd., Gateshead.

Mr. **G. H. Hodgson** has retired from his position of Director and Metallurgist with Kirkstall Forge Engineering Ltd., Leeds.

Mr. **J. R. Holland** has been awarded the degree of M.Met. in the University of Sheffield and has been appointed Research Associate in the Department of Metallurgy of the University of Kentucky, U.S.A.

Mr. **R. G. Hollister** has taken up a post as Metallurgist with Johnson Matthey and Co., Ltd., London.

Mr. **L. M. T. Hopkin** has been awarded the degree of Ph.D. in the University of London.

Mr. **A. D. Hopkins** has taken up an appointment with the Heat-treatment Service Department of the General Chemical Division of Imperial Chemical Industries Ltd., Oldbury.

Dr. **D. D. Howat** has left the Royal Technical College, Glasgow, to take up the position of Director of the Central Metallurgical Laboratory of the Anglo American Corporation of South Africa Ltd., Johannesburg.

Mr. **M. G. Huber** has left the Battelle Memorial Institute and has taken up an appointment as Metallurgical Engineer in the Special Steel Division of Union Carbide Europe, Geneva.

Mr. **H. P. Hughes** has been appointed General Works Manager and Technical Adviser of British Electro Metallurgical Company, Sheffield.

Mr. **G. J. T. Hume** has been awarded the degree of B.Sc. in the University of London and is undertaking post-graduate research at the Royal School of Mines, London.

Dr. **W. Hume-Rothery,** O.B.E., F.R.S., has been appointed George Kelly Reader in Metallurgy at Oxford University.

Mr. **V. L. James** has been appointed Lecturer in Metallurgy at Enfield Technical College, Middlesex.

Mr. **J. T. L. Jeffreys** has been awarded the degree of B.Sc. and is now on National Service in R.E.M.E.

Mr. **D. Johnston** has been awarded the degree of Ph.D. in the University of Sheffield.

Mr. **T. I. Jones** has taken up a post as Metallurgist with the Northern Aluminium Co., Ltd., Rogerstone, Mon.

Mr. **F. J. Kenaghan** has been awarded the degree of B.Sc. in the University of Birmingham and has been appointed Assistant Engineering Officer with the Atomic Weapons Research Establishment, Aldermaston, Berkshire.

Mr. **A. Knight** has been appointed Chemist with the U.K. Atomic Energy Authority, Aldermaston, Berkshire.

Dr. **E. B. Kula** has been appointed Physical Metallurgist with the Watertown Arsenal Ordnance Corporation, Massachusetts.

Mr. **J. D. Lancaster** has been awarded the degree of B.Sc. in Durham University and has taken up the post of Assistant Metallurgist with the Fairey Aviation Co., Ltd., Hayes, Middlesex.

Mr. **W. J. Lee-Bird** has taken up the post of Metallurgical Chemist with the Dunlop Rim and Wheel Co., Ltd., Coventry.

Mr. **B. A. Lloyd** has been awarded the degree of B.Sc. in London University.

Mr. **W. Lumb** has been appointed Metallurgist with Firth Vickers Stainless Steels, Sheffield.

Mr. **P. J. Macken** has been awarded the degrees of B.Sc. and A.R.S.M. in the University of London and has been appointed Assistant Metallurgist with the Copper Development Association, Radlett, Herts.

Mr. **D. Marles** has taken up the post of Works Assistant with Healey and Peart Ltd., Newport, Mon.

Mr. **J. W. Martin** has been awarded the degree of Ph.D. (Cantab.) and has become I.C.I. Fellow in the Department of Metallurgy, University of Cambridge.

Sir **Andrew McCance,** F.R.S., formerly Deputy Chairman and Joint Managing Director, has been appointed Chairman and Managing Director of Colvilles Ltd.

Mr. **T. C. Mead** has been awarded the degree of B.Sc.

Mr. **D. V. Mechem** has taken up an appointment with the Public Works Department Industrial Research Laboratories of the City of Birmingham.

Mr. **A. B. Mitchell** has been awarded the degree of B.Eng. in the University of Liverpool and has been appointed Assistant Experimental Officer with the U.K. Atomic Energy Authority, Culcheth.

Mr. **G. L. Moreton** has been appointed Superintendent of the Heat-treatment Service of the West Midlands Gas Board.

Mr. **A. R. Neelands** has been appointed Chairman of the Cementation Co., Ltd., London.

Dr. **L. Northcott** has been appointed Principal Superintendent, Materials Division, of the Ministry of Supply Armament Research and Development Establishment, Fort Halstead, Kent.

Mr. **D. G. Osborne** has been awarded the degree of B.Sc. in the University of London.

Mr. **K. D. Patten** is now a Pilot Officer in the Technical Branch of the R.A.F.

Mr. **A. J. Peech,** Deputy Managing Director of the

United Steel Companies, Ltd., has joined the Board of the Butterley Company Ltd.

Mr. **M. Perry** has been awarded the degrees of B.A. (Cantab.) and M.Met.(Sheffield) and has become George Senior Research Fellow in the Department of Metallurgy of the University of Sheffield.

Mr. **D. O. Pickman** has been appointed Superintendent of the Armament Research and Development Establishment, Fort Halstead, Kent.

Mr. **N. J. B. Pocock** has been awarded the degree of B.A. (Cantab.).

Mr. **W. L. Proctor** has been appointed Lecturer in Metallurgy and Fuel Technology at the Handsworth Technical College, Birmingham.

Mr. **S. V. Radcliffe** has been appointed University Research Fellow in the Department of Metallurgy of the University of Liverpool.

Dr. **E. G. Ramachandran** has been awarded the Kamani Gold Medal of the Indian Institute of Metals for 1955. The Medal is awarded annually for the best paper published in the Institute's *Transactions*.

Mr. **M. C. Regan** has been appointed Research Metallurgist with the Bristol Aeroplane Co., Ltd., Engines Division, Filton, Bristol.

Mr. **J. R. Richards** has been appointed Testing Assistant with Brown Bayley Steels Ltd., Sheffield.

Mr. **C. Roberts** has been appointed Physical Metallurgist with the Central Research Department of Dorman Long (Steel) Ltd., Middlesbrough.

Mr. **C. W. Sargeant** has been appointed Technical Representative with James Linday and Sons Ltd., Bilston, Staffordshire.

Mr. **A. G. Shakespeare** has been appointed Senior Scientific Officer with the South Wales Laboratories of B.I.S.R.A., Swansea.

Dr. **M. Signora** has been appointed Technical Director of Breda Siderurgica, Milan.

Dr. **I. G. Slater** has been appointed Head of the Department of Metallurgy at the College of Technology, Birmingham.

Mr. **E. S. Smith** has been appointed Technical Liaison Officer with Tube Investments Ltd., Birmingham.

Mr. **J. I. Spens** has been appointed Chairman of the Amalgamated Tin Mines of Nigeria Ltd.

Mr. **C. A. F. T. Spray** has taken up an appointment as Metallurgist with the Research Laboratories of the Mullard Valve Co., Ltd., Salford, Surrey.

Mr. **T. J. Szajda** has been awarded the degree of B.Sc. in the University of Wales and has taken up a post with the British Cast Iron Research Association, Birmingham.

Mr. **J. Thexton** has been appointed Chief Metallurgist of Henry Wiggin and Co., Ltd.

Mr. **W. K. Warburton** has relinquished his appointment with William Jessop and Sons Ltd., to take up a position as Research Metallurgist with the Stanton Ironworks Co., Ltd., near Nottingham.

Mr. **R. T. F. Waters** has left the English Electric Company to take up a post as Metallurgist with Standard Telephones and Cables Ltd., Rectifier Division, Harlow, Essex.

Obituary

Mr. **Harold Boot,** Managing Director of the Consett Iron Works and, since 1951, a Member of Council of the Institute, on 8th February, 1956.

Mr. **E. Cockin,** of Corby, Northants., in June, 1955.

Mr. **G. E. G. Gilfillan,** Special Director of the English Steel Corporation Ltd., and a Director of the English Steel Forge and Engineering Corporation Ltd., on 17th January, 1956.

Mr. **R. P. J. Paul**, on 12th January, 1956, following a motor-car accident.

Mr. **P. J. Rayner**, of Sheffield, on 5th September, 1955.

CONTRIBUTORS TO THE JOURNAL

G. Burns, M.Met., F.I.M.—Metallurgist with the Department of Engineering and Materials Research of the Admiralty.

Mr. Burns graduated from Sheffield University in 1924 as B.Met. and was subsequently awarded the degree of M.Met. in 1932. From 1924 to 1927 he was employed in the Metallurgical Laboratory at the Elswick Works of Sir W. G. Armstrong, Whitworth and Co., Ltd., Newcastle-on-Tyne, and from 1927 to 1934 in the Metallurgy Division of the Research Department, Woolwich. Papers on some of the work done during this time were published in the *Journal* in 1931 and 1932. From 1934 to 1938 he was with the Research Laboratory of the Mond Nickel Co., Ltd., Birmingham.

In 1938 Mr. Burns joined the Admiralty, serving until 1947 at the Metallurgy Section, Admiralty Engineering Laboratory, West Drayton, becoming Head of Section in 1945. From 1947 to 1949 he was Officer-in-Charge of the Central Metallurgical Laboratory, Emsworth. From 1949 to 1954 he was Head of the Metallurgy Division, Admiralty Materials Laboratory, Holton Heath. Since then he has served in the Department of Engineering and Materials Research and from mid-1955 has been attached to the Naval Section at A.E.R.E., Harwell.

Mr. Burns was elected a member of the Institute in 1927 and a Fellow of the Institution of Metallurgists in 1946.

E. J. Heeley, A.Met., F.I.M., A.M.I.Chem.E.—Chief Metallurgist of the Dyestuffs Division of Imperial Chemical Industries Ltd.

Mr. Heeley was educated at the Sheffield Central Secondary School and at Sheffield University, being awarded an Associateship in Ferrous Metallurgy in 1921. He served in the Research Department of Steel, Peech and Tozer Ltd., Sheffield, until 1928, when he joined the Vulcan Boiler and General Insurance Co., Ltd., Manchester, as Metallurgist. He took up his present appointment in 1941.

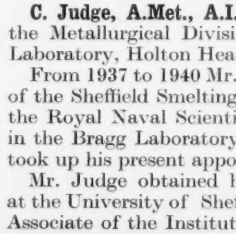
Mr. Heeley is a Fellow of the Institution of Metallurgists and an Associate Member of the Institution of Chemical Engineers.

C. Judge, A.Met., A.I.M.—Senior Scientific Officer in the Metallurgical Division of the Admiralty Materials Laboratory, Holton Heath.

From 1937 to 1940 Mr. Judge worked in the laboratory of the Sheffield Smelting Co., Ltd., Sheffield. He joined the Royal Naval Scientific Service in 1940 and worked in the Bragg Laboratory, Sheffield, until 1949, when he took up his present appointment.

Mr. Judge obtained his Associateship in Metallurgy at the University of Sheffield in 1946 and was elected an Associate of the Institution of Metallurgists in 1948.

Samuel J. Rosenberg—Metallurgist with the National Bureau of Standards, U.S. Department of Commerce, Washington, U.S.A.



Mr. Rosenberg was born in Washington, D.C., and graduated in mechanical engineering from the George Washington University in 1924. His entire professional career has been spent at the National Bureau of Standards, where he has been engaged in studies of various phases of physical metallurgy, with particular reference to steels.

A. T. Little, B.Sc. (Hons.)—Technical Officer in the Metallurgical Section of the Dyestuffs Division of Imperial Chemical Industries Ltd.

Mr. Little was educated at Kilmarnock Academy and at the University of Glasgow, where he graduated in 1939 with honours in chemistry. From then until 1944 he carried out research at Oxford University on the constitution of alloys, and took up his present appointment on leaving Oxford.

IRON AND STEEL ENGINEERS GROUP

The Thirtieth Meeting of the Iron and Steel Engineers Group will be held at the Offices of the Institute, 4 Grosvenor Gardens, London, S.W.1, on Thursday, 15th March, 1956. The Meeting will be devoted to the presentation and discussion of the paper on "*Modern Forging Presses and their Control*," by Mr. R. M. L. Elkan and Mr. J. T. Lewis, which was published in the February issue of the *Journal*, pp. 200–215.

The morning session, from 10.30 A.M. to 12.45 P.M., will be devoted to a discussion on Press Design and Performance; the afternoon session, from 2.0 to 4.0 P.M., to a discussion on Hydraulic Control Systems.

INSTITUTE OF METALS

Nuclear Energy Committee

The Council of the Institute of Metals has appointed a Nuclear Energy Committee with the object of stimulating interest in metallurgical problems arising in connection with the industrial and scientific aspects of the use of nuclear energy.

The Committee hopes to arrange an exhibition of nuclear energy products, of interest to metallurgists, in connection with the Institute's Spring Meeting in April, 1956.

Persons interested in the activities of the Committee should communicate with Lieut.-Colonel S. C. Guillan, Secretary, The Institute of Metals, 4 Grosvenor Gardens, London, S.W.1.

BRITISH IRON AND STEEL RESEARCH ASSOCIATION

Twenty-third Blast-furnace Conference

The 23rd Blast-furnace Conference was held at Harrogate on 19th and 20th October, 1955. Mr. R. A. Hacking, O.B.E., was in the chair.

The first session, on the disposal and utilization of slag, gave statistical data obtained from a recent questionnaire circulated to the industry by the British Iron and Steel Federation. The figures emphasized the large amount of slag that was tipped to waste.

In the first paper Mr. W. Banks gave a description of the disposal of slag as carried out by Colvilles Ltd., and Mr. C. W. Stubbins and Mr. D. A. Dowdeswell outlined the procedure followed by their respective firms. In the discussion Mr. L. G. Clugston and Mr. C. E. Hingley described the methods for preparing slag for commercial use. Dr. H. E. Z. Gordon also gave details of the Kinney–Osborne process for the production of expanded slag, and a film describing this process was shown later.

Dr. P. K. Gledhill introduced the second session with a paper describing the use of 100% sinter burden for a trial period on No. 1 furnace at Redcar. Mr. K. C. Sharp and Mr. J. G. Colvin noted that if burden preparation was carried out on a large scale, furnace operation would very quickly reflect any deterioration of quality in the prepared burden. Mr. J. Corke illustrated the difference in sulphur load on similar Redcar and Cleveland furnaces using a 100% sinter burden and a prepared burden, respectively. The principal benefits of sinter lay in the reduction of volatiles in the ore as charged to the furnace, although on one furnace it had been found that the degree of indirect reduction was less with a non-sintered burden than with the normal sintered burden. The general opinion was for increased sinter usage, but it was not thought that 100% would be ideal for all practices.

The first part of the third session was devoted to the use of high top pressure in the U.K. and the U.S.A. Mr. E. H. Baldwin gave details of the work carried out by his firm and Mr. A. T. Ledgard described the intensive use of high top pressure in U.S. blast-furnace plants. During the discussion it was pointed out that the steam consumption per ton of iron was the same in practice on a normal 28 ft 6 in. furnace as on an 18 ft 3 in. furnace with high top pressure.

A paper by Dr. F. G. Percival giving an account of the world's iron-ore resources concluded the last session. Some concern was expressed during the discussion on the low phosphorus content of some ores now being imported.

Staff Appointment

The Council of the British Iron and Steel Research Association has appointed Dr. J. Pearson as Assistant Director of Research, to assume such duties as are delegated to him by the Director, Sir Charles Goodeve. Dr. Pearson will continue his present duties as Head of the Steelmaking Division and Chemistry Department.

EDUCATION

Mond Nickel Fellowships

The Mond Nickel Fellowships Committee invites applications for five Fellowships of an approximate value of £900 to £1200 each for 1956. Fellowships will be awarded to selected candidates of British nationality with degree or equivalent qualifications to enable them to obtain wider experience and additional training in industrial establishments, at home or abroad, to make them more suitable for future employment in senior technical and administrative positions in British metallurgical industries. Each Fellowship will cover one full working year. Applicants will be required to state details of the programme they wish to carry out. Particulars and forms of application are available from: The Secretary, Mond Nickel Fellowships Committee, 4 Grosvenor Gardens, London, S.W.1. Completed application forms are required by 1st June, 1956.

NEWS OF SCIENCE AND INDUSTRY

Symposia on Corrosion—New York

The National Association of Corrosion Engineers is holding its twelfth annual Conference and Corrosion Show in New York City from 12th to 16th March, 1956. Papers covering a wide variety of corrosion problems met in steel metallurgy and at elevated temperatures will be given.

I.E.E. Utilization Section Meeting—London

At the meeting of the Utilization Section of the Institution of Electrical Engineers to be held at Savoy Place, London, W.C.2, at 5.30 P.M. on 19th April, 1956, the following papers will be presented: "*Temperature Rises in Electrical Machines as Related to the Properties of Thermal Networks*," by Dr. J. J. Bates and Professor A. Tustin; "*Temperature Rises in Electrical Machines on Variable Load with Variable Speed*," by Professor A. Tustin and Dr. J. J. Bates; "*Performance and Heating Curves for Motors on Short-run Duties*," by Professor A. Tustin, Mr. D. F. Nettell, and Mr. R. Solt. The Institution have extended an invitation to members of the Institute's Engineers Group to attend the meeting and take part in the discussion.

DIARY

1st Mar.—Institute of Metal Finishing (North-West Centre)—"*General Plating Shop Problems and suggested Remedies*," by R. G. Hughes—The Engineers' Club, Albert Square, Manchester, 7.30 p.m.

1st Mar.—Leeds Metallurgical Society—"*Extrusion*," by J. Crowther—Large Chemistry Lecture Theatre, University of Leeds, 7.15 p.m.

1st Mar.—Liverpool Metallurgical Society—Conversazione—The Liverpool Engineering Society, The Temple, Dale Street, Liverpool. 7.00 p.m.

2nd Mar.—Society of Chemical Industry—"*Corrosion Control by Water Treatment*," by R. S. Thornhill. "*Water Treatment for Recirculating Water Systems*," by C. W. Drane. "*Boiler Water Conditioning at Ultra-high Pressures*," by C. D. Weir. (Joint meeting with Stirlingshire Section)—Grangemouth, 2.30 and 7.30 p.m.

5th Mar.—Cleveland Institution of Engineers—"*Engineering Aspects of Heavy Section Mills*," by T. Walker—Cleveland Scientific and Technical Institution, Corporation Road, Middlesbrough, 6.30 p.m.

5th–7th Mar.—Chemical Institute of Canada, Chemical Engineering Division—Sixth Conference—Sarnia, Ontario.

6th Mar.—Sheffield Metallurgical Association—"*Some Recent Developments in Metallurgical Analysis*," by G. W. C. Milner—B.I.S.R.A. Laboratories, Hoyle Street, Sheffield 3, 7.00 p.m.

8th Mar.—Liverpool Metallurgical Society—"*Recent Developments in Casting and Moulding Techniques*," by A. H. Sully—Liverpool Engineering Society, The Temple, Dale Street, Liverpool, 7.00 p.m.

8th Mar.—Staffordshire Iron and Steel Institute—"*The Recovery of Spent Pickle Liquor*," by W. Bullough, 7.30 p.m.

12th Mar.—Institution of Works Managers (Manchester Branch)—"*Industrial Engineering—Its Scope and Opportunities*," by W. J. Worsdale—Grand Hotel, Manchester, 6.45 p.m.

12th–16th Mar.—National Association of Corrosion Engineers—Twelfth Annual Conference and Corrosion Show—Statler Hotel, New York City.

13th Mar.—Institution of Engineering Inspection, Leeds—"*Principles of Mechanical and Metallurgical Testing Methods of Materials*," by J. W. Poole—Theological Library, Leeds Church Institute, Albion Place, Leeds 1, 7.30 p.m.

13th Mar.—North East Metallurgical Society—"*An Outline of the Modern Metallurgy of Cast Iron*," by H. Morrogh—Cleveland Scientific and Technical Institution, 7.15 p.m.

13th Mar.—Sheffield Metallurgical Association—Biennial Dinner at the Royal Victoria Hotel.

14th Mar.—East Midlands Metallurgical Society—"*Production, Fabrication, and Properties of Titanium and Some Titanium Alloys*," by N. P. Inglis—Nottingham and District Technical College, Shakespeare Street, Nottingham, 7.30 p.m.

14th Mar.—Liverpool Metallurgical Society—Visit to Leyland Motors Ltd., Leyland, Lancashire.

14th Mar.—Manchester Metallurgical Society—"*Some Metallurgical Problems Imposed by Stratospheric Flight*," by Major P. Litherland Teed—Lecture Room, Central Library, Manchester, 6.30 p.m.

14th Mar.—Newport and District Metallurgical Society—"*The Electric Furnace in the Steel Industry*," by J. C. Howard—Whitehead Institute, Cardiff Road, Newport, 7.00 p.m.

16th Mar.—West of Scotland Iron and Steel Institute—"*Recent Investigations into the Chemistry of Open-hearth Roof Refractories and their Bearing on Performance*," by J. White—39 Elmbank Crescent, Glasgow, 6.45 p.m.

17th Mar.—Swansea and District Metallurgical Society—General Meeting, "*Modern Developments in Tinplate Manufacture*," by J. S. Gazard—Central Library, Swansea, 6.30 p.m.

19th Mar.—Sheffield Society of Engineers and Metallurgists—A paper on the 60-ton Arc Furnace at Stocksbridge—The University Building, St. George's Square, Sheffield, 7.30 p.m.

20th Mar.—Institution of Naval Architects—Annual Dinner—Grosvenor House, Park Lane, London, W.1, 7.00 for 7.30 p.m.

20th Mar.—North East Metallurgical Society—"*Powder Metallurgy*," by I. Jenkins—Cleveland Scientific and Technical Institution, Middlesbrough, 7.15 p.m.

20th Mar.—Sheffield Metallurgical Association—"*Some Problems Associated with the Production of Heavy Forgings*," by T. R. Middleton—B.I.S.R.A. Laboratories, Hoyle Street, Sheffield 3, 7.00 p.m.

21st Mar.—Institute of Fuel (North Western Section)—"*Steam Raising from Waste Heat*," by C. Ungoed—The Engineers' Club, Albert Square, Manchester, 2.30 p.m.

21st Mar.—North Wales Metallurgical Society—"*Fuels in an Integrated Steelworks*," by A. D. Osborne—County Primary School, Plymouth Street, Shotton, Near Chester, 7.15 p.m.

22nd Mar.—Ebbw Vale Metallurgical Society—Joint Meeting with The Iron and Steel Institute—Speaker, Sir Charles Goodeve, F.R.S.

22nd Mar.—Institution of Engineering Inspection (North Western Branch)—"*The Application of Ultrasonics to Engineering Inspection*," by J. Brigg—The Engineers' Club, Albert Square, Manchester, 7.30 p.m.

22nd–23rd Mar.—Institution of Mining and Metallurgy—Symposium on "*The Extraction Metallurgy of Some of the Less Common Metals*"—Lecture Hall, Royal Society of Arts, 6 John Adam Street, London, W.C.2.

23rd Mar.—Ebbw Vale Metallurgical Society—Annual Dinner.

23rd Mar.—Manchester Association of Engineers—Annual General Meeting, and "*The Production of Steel Strip and Sheets*," by H. Edwards—The Engineers' Club, Albert Square, Manchester, 6.45 p.m.

24th Mar.—Swansea and District Metallurgical Society—Annual Dinner—Central Library, Swansea, 6.30 p.m.

27th Mar.—Sheffield Metallurgical Association—"*Monolithic Linings*," by W. H. Everard—B.I.S.R.A. Laboratories, Hoyle Street, Sheffield 3, 7.00 p.m.

TRANSLATION SERVICE

(The previous announcement was made in the February, 1956 issue of the *Journal*, p. 218).

Translation Available

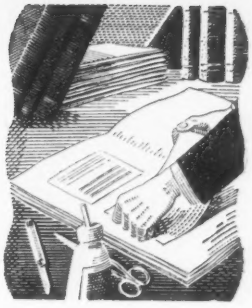
No. 508 (Italian). V. Ferrari and D. Bellini: "Horizontal Pouring." (*Met. Ital.*, 1954, vol. 46, Nov., pp. 417–420).

Translation in Course of Preparation

(German). W. Schmitt: "Problems of the Efficiency of Open-hearth Furnaces using the Pig-iron/Scrap Process." (*Stahl u. Eisen*, 1955, vol. 75, 1st Dec., pp. 1627–1640).

ABSTRACTS OF CURRENT LITERATURE AND BOOK NOTICES

Iron and Steel Manufacture and Related Subjects



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PRODUCTION OF STEEL

Contributions to Improvements of Steel Quality. W. Moll. (*Met. u. Giesserei Techn.*, 1951, **1**, Sept., 261–265; Oct., 307–309). A historical review of developments in the concept of quality of steels is given with special reference to those factors that are considered to have made progress possible. The present state of knowledge is surveyed, with special reference to the influence of oxygen on steel properties.

Problems of Scrap Supply in Steel Works. W. Dams. (*Met. u. Giesserei Techn.*, 1951, **1**, Aug., 253–254; Sept., 266–267; Nov., 339–340). The author deals first with a method of investigating the problem under the headings of preparation, execution, evaluation, introduction, control and results. In the second part the problems of scrap handling and transport are considered and finally the system introduced at the Ricsa Steel Works, which enabled the turn-round of scrap hoppers to be speeded up by half, is explained.—L. J. L.

Possible Ways of Increasing the Life of Open Hearth Furnaces and of Improving their Performance. P. P. Budnikow, D. P. Bogatzki, and J. L. Rosenblit. (*Met. u. Giesserei Techn.*, 1951, **1**, Nov., 325–330; *Proc. Acad. Sci.* (Moscow), 1951, (6), 901–913). Suggestions for the rational use of highly heat resistant materials in furnaces are made; the replacement, by these materials, of Dinas stone is advocated, design and erection of basic scaffolds in open hearth furnaces are discussed; and the construction of open hearth furnaces of completely basic materials is described.—L. J. L.

On the Fundamental Conditions of the Steelmaking Process by Oxygen Blowing. Y. Katsufuji and K. Niwa. (*Tetsu to Hagane*, 1955, **41**, May, 500–505). [In Japanese]. C removal occurs at the gas/liquid interface, and its rate depends on the weight of Fe, the area of interface and the O_2 pressure at a given C content. In practice the interface area varies with the bubbles (containing O_2 at approx. 1 atm) produced, and hence on the amount of O_2 introduced. The diffusion of C in molten Fe should be aided by physical turbulence. O_2 blowing is more effective at low C concentration.—K. E. J.

Interfacial Tension between Molten Iron and Slags. K. Mori and T. Fujimura (*Tetsu to Hagane*, 1955, **41**, May, 495–500). [In Japanese]. Values of interfacial tension were determined in an alundum crucible in which slag was melted, by photographing the sessile Fe drop by X-ray, calculating its dimensions, and using the Bashforth–Adams tables. For liquid Fe–Si alloy (14·34% Si) and Na_2O–SiO_2 and Li_2O–SiO_2 slags at 1370° C, the values were approx. 1000–1150 dyne/cm, decreasing with increase of alkaline oxide. Energies of adhesion were approx. 500–700 erg/cm², values agreeing with earlier work on metal–metal-oxide systems. (19 references).—K. E. J.

Decarburization in the Basic Open-Hearth. M. Urbain and P. Flament. (*Rev. Mét.*, 1955, **52**, Mar., 179–189). By calculating the partial pressures of carbon monoxide, it is shown that these pressures are lower for the smaller rates of decarburization. The means of selecting the carbon content of the charge in order to finish the operation with desired carbon content and with the minimum amount of oxygen is described. Increase in the nitrogen and hydrogen contents may be achieved by heavy additions of ferro-silicon.—G. E. D.

The Pre-Refining of Pig-Iron in the Ladle by Means of a Pure Oxygen. Results obtained in a Pilot Plant. P. Leroy and L. Septier. (*Centre Doc. Sidér., Circ. Inform. Tech.*, 1955, (7), 1383–1402). The advantages of the pre-refining of pig-iron in the ladle are discussed, and different methods of introducing oxygen are described. These include oblique blowing during filling, and high-pressure blowing. Details of the Mondeville pilot plant and iron oxide fume precipitator are given. Results of a 10 day trial are reported.—T. E. D.

Desulphurization by Agitation with Slag. R. Perrin. (*Rev. Mét.*, 1955, **52**, June, 473–476). Methods of steel desulphurization are broadly discussed, with particular consideration of the method of stirring with molten slags. The principles are also given of a new method of simultaneous desiliconization and desulphurization of steel, using a single slag of appropriate composition.—G. E. D.

Separation of Soda Slag from Hot Metals: Works Observations and Laboratory Experiments. C. E. A. Shanahan. (*Iron Steel*, 1955, **28**, Apr., 123–127). A serious difficulty of the soda-arch process is the removal of the soda slag from the hot metal. The author describes works observations and laboratory experiments on the rate of reparation of the slag by flotation into a surface layer and on methods of separating this surface slag phase.—G. F.

B.O.C. in Industry. Steelmaking. (*Torch*, 1955, (5), 2–10).

Some Problems Relating to the Construction of Arc Furnaces. K. Rosenzweig. (*Hutník* (Prague), 1955, **5**, (7), 197–202). [In Czech]. A Czech arc furnace of the rotating-roof type is described, and the designs of arc furnaces produced in most steel producing countries are assessed. The Swedish

ASEA and the recently designed Czech furnaces are considered best on the basis of a points system of assessment; Swiss and British furnaces follow.—P. F.

Electromagnetic Stirring. Its Application to Desulphurization. J. Duflot and M. Porcheray. (*Rev. Mét.*, 1955, **52**, Apr., 335–347). The results of tests using an electro-magnetic stirrer in the electric furnace at the Hagondage works of the Société des Aciers Fins de l'Est are discussed. The principal advantages claimed are increased desulphurization, more rapid homogenization of the bath, easier stirring and the improved quality of certain types of steel.—G. E. D.

A Method of Repair for the Hearths of Electric Arc Furnaces. W. Dröge. (*Met. u. Giesserei Techn.*, 1951, **1**, Nov., 340–341). In order to repair craters forming under the electrodes in electric furnaces, a groove is burned with oxygen to the lowest point of the crater immediately after the furnace has been emptied, steel is drained through the groove, the furnace is righted, the crater is filled with dolomite which is rammed down by means of a weight of 2 tons.—L. J. L.

Hydrogen in Molten Steel. I. Behaviour of Hydrogen in Molten Steel in Electric Furnace Steel Making Process. S. Sawa. (*Tetsu to Hagane*, 1955, **41**, Apr., 416–424). [In Japanese]. The effects of H_2O in the atmosphere, slag conditions, C content of the steel, melting conditions, O_2 content of the steel, etc., were investigated, using statistical methods, in basic and acid electric furnaces. The H_2 content tended to reach a steady value, according to the H_2O content in the furnace atmosphere; this was greatest under basic reducing slag conditions.—K. E. J.

On a New Procedure for the Treatment of Metals Using Solar Energy. F. Trombe and M. Foex. (*Usine Nouvelle*, 1955, **11**, Mar. 17, 62). A rotating furnace, with or without an inlet for gas, heated by a 75 kW solar furnace, is briefly described. The method by which a refractory crucible is produced within it is described. A mass of 80 kg of iron can be melted in this and submitted to any required chemical or alloying treatment.—T. E. D.

Install Vacuum Setup for Dual Melting Role. W. G. Patton. (*Iron Age*, 1955, **175**, June 23, 95–97). Applications for vacuum melted alloys are noted and a new high capacity plant at General Electric Co. is described. 1000 lb melts can be made and cast. The oil diffusion pump can take pressures down to 1 μ.—D. L. C. P.

Some Aspects of the Thermodynamic Properties of Slags. J. E. Antill and P. Murray. (*Silicates Indust.*, 1955, **20**, Aug.-Sept., 293–298). Phase boundaries in binary oxide and fluoride systems are determined from thermodynamic principles, and the extension of the methods to ternary systems is indicated. Applications of the theory to refractories are discussed.—P. F.

What are the Reactions that Cause Frothing of Slag in the Basic Open Hearth Process? K. Fellcht. (*Metallurgie*, 1955, **5**, Mar., 85–102). The author concludes that frothing of slag is due to high surface tension of the slag combined with simultaneous and violent generation of CO.—L. J. L.

Study on Ingot-making Practice. V. S. Maekawa and Y. Nakagawa. (*Tetsu to Hagane*, 1955, **41**, June, 581–587). [In Japanese]. The degree of oxidation of steel and the production of non-metallic inclusions during the pouring of steel from the ladle to the mould were investigated quantitatively under various conditions. More oxidation is found with a turbulent metal stream. Si, Mn and Fe in a high-C steel were not oxidized to such an extent as in other steels.

On Several Improvements Related to The Continuous Casting of Light and Heavy Metals and Alloys. G. Befayt. (*Usine Nouvelle*, 1955, **11**, May 26, 31–33). Continuous casting of iron, copper, aluminium and their alloys is discussed, and suggestions are made. Points examined are the cooling system, the means of withdrawing the ingot, and direct rolling—T. E. D.

Continuous Casting of Steel at the "Anciens Etablissements Cail" at Denain (Nord). (*Usine Nouvelle*, 1955, **11**, May 12, 31–32). The installation was put into use in Dec. 1954. A brief history of continuous casting is given, and overseas plants are mentioned. The Cail plant has a capacity of 45 tons/h and has been used for a variety of steels. Some details of plant practice, quality of the products, and output are given.—T. E. D.

A Special Type of Ingot Crack Caused by Certain Impurities. H. Björnson and H. Nathorst. (*Jernkontorets Ann.*, 1955, **139**, (6), 412–438). [In Swedish]. Details are given of cracks formed in ingots when the steel contains copper and tin. Cu and Sn segregate to the primary grain boundaries on cooling, thus forming the precipitate of probably the liquid Cu–Sn phase. The mechanism of crack formation is discussed, and the three main factors considered are: 1, reduction in cohesive strength; 2, brittleness of duplex structures where one of the constituents forms thin films round the other; 3, stresses caused by the temperature gradient.—G. G. K.

FOUNDRY PRACTICE

Historical Exhibition of Foundry Science. O. Onorato. (*Fonderia Ital.*, 1955, **4**, June, 303–312). [In Italian]. The author gives a detailed and learned account of the development of foundry art and science. (37 references).—M. D. J. B.

Old and Modern Process in Foundries. G. Grosser. (*Fonderia Ital.*, 1955, **4**, May, 248–250). [In Italian]. The author discusses briefly the salient features of a modern foundry layout stressing the importance of good materials flow lines.

Progress Made in the Foundry in 1954. P. A. Heller. (*Stahl u. Eisen*, 1955, **75**, July 14, 914–918; July 28, 990–993). A review of published information on foundry practice. In the first part the author deals with structural—mainly graphite formation including spheroidal graphite—and mechanical properties. In the second part, melting practice, moulding and casting, standardization, flowability and electrolytic polishing of grey iron are discussed.—T. G.

Some Causes of Wasters in Grey Iron Casting. F. Barheine. (*Met. u. Giesserei Techn.*, 1951, **1**, May, 134–135). Methods of remedying piping are listed as follows: (1) reduction of gas involved by adequate drying of cores; (2) changes of design, such as increase in number of rents; (3) quicker casting; (4) use of ingot moulds.—L. J. L.

Use of High-Quality Pig Iron as a Foundry Production Reserve. L. I. Gol'denberg. (*Liteĭnoe Proizvodstvo*, 1955, (3), 7–9). [In Russian]. The effect of using various pig irons in the cupola charge is discussed in relation to their structure, composition and other properties.—S. K.

Casting. (*Iron Age*, 1955, **175**, June, B2–B14). Past, present and future trends in casting practice are reviewed. Modern moulding techniques are surveyed and an increase is predicted in ductile iron applications and gray iron casting capacity by the shell process.—D. L. C. P.

On Modifications in Practice and Administrative Organization in Foundries. R. Teychenné. (*Usine Nouvelle*, 1955, June, 4–9). The way in which problems of practice and administration can be tackled by experts, with a view to improving production, is described.—T. E. D.

Contribution to the Nomenclature of Casting Material and Production Castings. W. Patterson. (*Giesserei*, 1955, **42**, June 23, 333–337). The systematic arrangement of castings is considered from the following viewpoints: (a) alloy compositions (b) casting methods, (c) mould material used, (d) patterns used, and (e) manufacturing methods. The systematic classification of casting alloys is then described, first in two groups, ferrous and non-ferrous, then divided and subdivided. Proposals are made for a shortened nomenclature of cast alloys which shows the % content of the principal constituents.—R. J. W.

External Water Cooling of Cupolas. G. Östberg. (*Gjuteriet*, 1955, **45**, June, 82–83). [In Swedish]. Details are given of Hubert and Danis's experiences when applying water spray cooling to the exteriors of cupolas with 820, 1000, and 1500 mm dia., also information regarding safety measures, resultant heat losses, and charge temperature. Flaking of the lining was countered by reducing the degree of ramming. The amount of coke employed had to be increased by 1% however, to compensate for the heat required to raise the cooling water temperature to 60° C.—G. G. K.

Designing a Small Cupola. C. W. Ammen. (*Foundry*, 1955, **83**, June, 128, 131, 134, 137, 140).

Characteristics Determining the Use of Rammed Linings. (*Fonderie*, 1955, Mar., 4435–4437). The properties of refractories that should be considered in choosing a suitable furnace lining are briefly discussed. The correct method of lining a cupola is also outlined.—B. C. W.

Blast Pre-Heating in the Cupola. G. Sper. (*Met. u. Giesserei Techn.*, 1951, **1**, May, 136–138). The Schürmann, Hörnig and Piwowarsky systems of air-blast preheating are briefly described.—L. J. L.

Effects of Blast Humidity on Iron Properties. J. L. Brooks. (*Foundry*, 1955, **83**, May, 219, 222). The effects of air blast humidity on cupola operation are outlined and the use of an

absorption type dryer to maintain constant moisture content is briefly described.—B. G. W.

The Use of Briquettes from Steel Turnings in Foundry Practice. F. Naumann. (*Met. u. Giesserei Techn.*, 1951, **1**, Mar., 85–86). A method of making briquettes from steel shavings and turnings for use in the cupola is briefly described and illustrated.—L. J. L.

Heat Economy in the Cupola. W. Callenberg. (*Met. u. Giesserei Techn.*, 1951, **1**, July, 198–205). Coke quality, coke consumption, measurement of blast volume and blast pressure, control of furnace gas composition and temperature, and measurement of casting temperature are discussed.

Research on the Relation between Cast Iron and Oxygen: on the Effect of Oxygen on the Mechanical Properties of Cast Iron and on the Classification of Cast Iron by the Difference of Oxygen Content in Cast Iron. M. Homma and M. Shigihara. (*Nippon Kinzoku Gakkai-Si*, 1952, **16**, Dec., 666–670). [In Japanese]. Increase in the O_2 content of flake-graphite iron led to increase of tensile strength and lowered toughness; reduction of O_2 content produced stronger, tougher irons. Changes in properties caused by O_2 variations are discussed with reference to melting practice and to variations in properties of irons of identical composition and thermal history. (18 references).—K. E. J.

Production of High-Strength Cast Iron in the Cupola Forehearth. I. P. Petrov. (*Liteĭnoe Proizvodstvo*, 1955, (1), 3–5). [In Russian]. This is an account of the inoculation of cast iron with magnesium carried out in the forehearth of a hot blast cupola producing 7–8 tons/h. Normal grey iron is produced until the forehearth is fully heated, as indicated by the temperature stability of the iron flowing out. Slag is then run out and magnesium is introduced below the metal surface with a special steel bell. Any magnesium fumes produced pass into the cupola. This procedure is said to minimise the hazards of inoculation and give a very uniform iron with excellent mechanical properties.—S. K.

Influence of the Chemical Composition of Cast Iron on the Resistance to Wear of Brake Shoes. D. T. Zelenskiĭ. (*Liteĭnoe Proizvodstvo*, 1955, (2), 7–10). [In Russian]. This article deals with an experimental investigation of the effect of chemical composition and other factors on the life of railway wagon brake shoes. Experimental batches were cast in sand, metal and cermet moulds using irons with the following composition-range: C, 3·13%–3·5%, Si, 0·71–1·60%, Mn, 0·79–1·44%, P, 0·26–0·38%, S, 0·059–0·10%. These shoes were subjected to metallographic investigation and service tests. Further test-pieces were cast in thin-walled steel moulds with chilling. Optimal conditions for brake-shoe casting are given together with the design of a special mould.—S. K.

Pearlitic Cast Iron of High Carbon and Sulphur Content. G. Sper. (*Met. u. Giesserei Techn.*, 1951, **1**, Aug., 231–238). The author gives a short historical survey and then deals with analyses of near-eutectic cast iron, graphite crystallization, casting techniques, non-metallic nuclei, critical Si content, graphite distribution, freezing, manganese sulphide, undercooling, new processes, and iron- and slag-analyses.—L. J. L.

The Production of High Quality Cast Iron in the Electric Furnace. A. Lincke. (*Met. u. Giesserei Techn.*, 1951, **1**, Sept., 270–273). The author claims that the duplex process described has shown that, apart from carbon, silicon, phosphorus, and sulphur, the superheat temperature of the melt will have the greatest effect on the physical properties of cast iron, and that superheating alone will determine the quality of the product.—L. J. L.

The Effect of Inoculation in Nodular Cast Iron. I. Iitaka and K. Nakamura. (*Waseda Univ., Rep. Castings Res. Lab.*, 1955, (6), 9–11). Tests using Fe–Si and Ca–Si as inoculants are reported. The effect of inoculation, in nodular iron, is the same as that of inoculants of graphitization in normal flaky-graphite iron. The specific inoculants inhibit the tendency of Mg to produce a white iron, and influence the matrix structure, and size and number of graphite particles, but not the shape of the last.—K. E. J.

The Sulphurization-Resistant Property of Spheroidal Graphite Cast Iron at High Temperatures. M. Shiozawa and H. Nakai. (*Waseda Univ., Rep. Castings Res. Lab.*, 1955, (6), 12–14). Tests were made with molten Swedish and Australian irons, with additions of Mg (from Fe–Si–Mg) and Al. The resistance to sulphurization of nodular iron is not different from that of ordinary cast iron. The shape of the graphite exerts no influence, but the resistance to sulphurization is seriously affected by the alloying elements dissolved in the matrix.

The Effects of Mo, Ni: Mo, and Cu: Mo Additions on the Matrix of Spheroidal Graphite Cast Iron. T. Kusakawa. (*Waseda Univ., Rep. Castings Res. Lab.*, 1955, (6), 15–20). Additions of Mo, Ni–Mo or Cu–Mo may produce a supercooled matrix. Subsequent treatment with Mg leads to a tendency towards bainite formation: a higher hardness (by 30–40 Rockwell C) is also possible. Without Mg additions, the hardness is not so responsive to Si additions as with Mg additions when increase of Si content decreases the hardness.—K. E. J.

On Various Theories of Globular Graphite Formation in Cast Iron. I. Iitaka. (*Waseda Univ., Rep. Castings Res. Lab.*, 1955, (6), 1–5). Existing theories of spheroidal graphite formation are reviewed. A new theory is advanced, based on the facts that all shapes of graphite found in ordinary Fe have their counterparts in natural graphite, and that graphite is flexible and inelastic at room temperature. The formation of nodular graphite results from the absorption by graphite particles of Mg vapour released from solution on cooling, which decreases surface tension and renders the particles deformable. (22 references).—K. E. J.

Special Features of the Production of Magnesium Cast Iron. V. A. Zakharov. (*Liteĭnoe Proizvodstvo*, 1955, (2), 14–16). [In Russian]. Experiments on the dilution of magnesium-treated cast iron with ordinary cast iron and on the change in the carbon-content of cast iron after addition of magnesium are described. Dilution with ordinary iron failed to prevent decarburization of the castings but caused the amount of graphite spheroids to increase by 10–15%. To study the behaviour of carbon, specimens of magnesium-inoculated iron were cast into sand and chill moulds, with and without ferro-silicon additions. The analyses, fractures and micro-structures were compared.—S. K.

Causes and Prevention of Peripheral Pearlite in Black Heart Malleable Cast Iron. A. Bordes. (*Fonderia Ital.*, 1955, **4**, June, 277–295). [In Italian]. After referring briefly to the effects of peripheral pearlite on the mechanical properties of cast iron, the author examines in detail the various factors which may be responsible for its formation. Practical tests are described which were carried out to determine the influence of the cupola change, the method of smelting, the thickness and shape of the test piece and the nature of the heat treatment. Considerable technical information and a large number of micro-photographs are given.—M. D. J. B.

On Austenitic Malleable Iron. II. N. Tsutsumi. (*Waseda Univ., Rep. Castings Res. Lab.*, 1955, (6), 21–26). Results are given for tests of mechanical properties of a wide range of irons treated with Cu, Ni and Al, and for the increases produced by water-quenching. Details of structures are given. Ni additions can replace Mn to confer carbide-stabilizing and austenite-forming properties. Reducing or carburizing atmospheres are necessary for the manufacture of these irons, but close degrees of temperature control are not necessary. The irons will probably replace Mn steel or pearlitic blackheart malleable irons for applications requiring high wear- or corrosion-resistance.—K. E. J.

Quality-Control of Anti-Friction Malleable Cast Iron in Mass-Production Conditions. Ya. G. Lifshits. (*Liteĭnoe Proizvodstvo*, 1955, (1), 26). [In Russian]. The results of a statistical treatment of two years' test data on the hardness, pearlite-content, temper-carbon content, and grain size of pearlitic-ferritic malleable cast iron used as a substitute for bronze in bearings, are given. On the basis of this investigation it has been found possible to modify advantageously the routine test procedure: only 3% of the output is now subjected to micro-structure study, 100% being hardness tested. The micro-structure is checked only to ensure that lamellar graphite, cementite and granular pearlite are absent.—S. K.

Precision Alloy Steel Castings. (*Machinery Lloyd, European Ed.*, 1955, **27**, May 28, 49–52).

Some New Trends in the Converter Process for Production of Steel Castings. S. Pelczarski and T. Heinar. (*Przeglad Odlewnictwa*, 1955, 5, (3), 61–73; 1955, **5**, (4), 97–102). [In Polish]. The historical development of the process, its metallurgy and the properties of steel produced are described. Side blown converter and L. D. practices and their advantages and disadvantages are discussed.—V. G.

On the Melting and Thermal Cracking of Austenitic Manganese Steel Castings. F. C. Althof. (*Giesserei*, 1955, **42**, July 7, 362–370). The melting of manganese steel (11–14% Mn)

in a basic electric arc furnace is first described in detail. The formation of intercrystalline thermal cracks is described and illustrated before and after a homogenizing treatment. The occurrence of thermal cracks is described during the manufacture of chain links and its dependence on phosphorus content established. The influence of moulding material composition on the heat cracking of these chain links has also been investigated systematically. The mechanism of crack formation is discussed.—R. J. W.

The Effect of the Metallurgical Properties of Steel Castings on the Formation of Hot Tears. H. Heyer and E. Piwowarsky. (*Giesserei*, 1955, **42**, May 26, 273–279). The terminology and diagnosis of thermal cracking is first discussed. The factors affecting heat crack formation are (a) the metallurgical state of the steel, and (b) external factors. The design of the apparatus used for the heat crack experiments is illustrated and described. The effect of casting temperature, the grain size and the carbon, phosphorus and sulphur contents of unalloyed cast steel on the formation of heat cracks was investigated. The results are presented and analysed statistically. The metallurgical properties were found to have an effect on heat crack formation: the kind of solidification undergone by the casting was also important.—R. J. W.

Method to Determine the Semblance of a Moulding Sand Grain. J. Kashima. (*Waseda Univ., Rep. Castings Res. Lab.*, 1955, (6), 31–37). A method is described by which the specific surface area of sands, in terms of the grain shape factor, is determined by the amount of ethylene glycol or glycerine remaining per unit weight after continued centrifuging compared with that remaining on unit weight of artificially produced spherical glass grains. 30–40 min are required for the determination, and the accuracy is within 5%.

X-Ray Analysis of Microstructure and its Application to the Examination of Mould and Auxiliary Materials in Foundry Practice. M. Weidner. (*Giesserei Technik*, 1955, **1**, May, 64–68).

The Making of Patterns by the Basket System. H. Klotz. (*Giesserei Technik*, 1955, **1**, May, 73–74). The patterns described consist of a conical basket, to which external walls are fitted in sections.—L. J. L.

Feeding and Solidification: The Position and Size of Feeders. K. Åkesson. (*Gjuteriet*, 1955, **45**, Mar., 31–38). [In Swedish]. The position of the riser in relation to the heat centre of the casting is discussed and graphical and mathematical methods for determining riser size are described. A new equation, based on Chworinoff's rule, is presented for the calculation of risers.—G. G. K.

Automatic Moulding Machines. J. Broberg, L. B. Lindh and K. Åkesson. (*Gjuteriet*, 1955, **45**, May, 61–65). [In Swedish]. After some general views on automatic moulding aspects such as necessary patterns, moulding materials, mould sections and machine maintenance, the author gives brief descriptions of four Swedish machines, three of the conventional type and one with circulating pattern plates.

Moisture Absorption in Dried Cores. R. Moren and S. Bergmann. (*Gjuteriet*, 1955, **45**, Apr., 51–54). [In Swedish]. Details are given of a laboratory method for studying water absorption under controlled conditions. Tests to date show that there are marked differences in the rates of water absorption encountered when different binders are employed to impart green strength to an oil-sand core. Dextrin caused the highest degree of absorption whilst minimum absorption was obtained with cellulose derivatives.—G. G. K.

The Economic Drying of Moulds and Cores. W. Callenberg. (*Met. u. Giesserei Techn.*, 1951, **1**, May, 139–149). Recent developments are not discussed in detail, but it is pointed out that drying by infra-red radiation had not been found satisfactory. Suggestions are made to enable foundries to modernize their drying plant—L. J. L.

The CO_2 Process—Its Use in British Foundries. A. Tipper. (*Foundry*, 1955, **83**, June, 84–90). In the CO_2 process for hardening moulds or cores the gas is blown through dry clay-free silica sand to precipitate silica in the form of a gel from the sodium silicate binder. The preparation of cores and moulds using this process is outlined and examples of its successful applications are given. Methods of applying the CO_2 are reviewed and the advantages and disadvantages of the process are summarized.—B. C. W.

Self-Drying Core Binders. P. Nicolas. (*Fonderie*, 1955, Apr., 4474–4482). Data are presented to show the effect of various factors on the setting of self-drying core binders. Among the factors considered are the time of mixing, the grain size of the sand, the curing temperature and the presence of clay, silica flour and organic matter in the sand. Difficulties occasionally experienced with the binders are briefly discussed.—B. C. W.

The Design of Castings in Grey Cast Iron. (*Usine Nouvelle*, Special No., Spring, 1955, 25–33). Illustrated examples are given of correct and incorrect castings design. Where possible, tensile stresses should be avoided. Mechanical properties of various sections are compared. Design for resistance to shock, and to avoid notch embrittlement is mentioned. Profiles of castings for convenient machining are shown and discussed.—T. E. D.

Scale Formation in the Production of Ingot Moulds. W. Brenner. (*Met. u. Giesserei Techn.*, 1951, **1**, May, 152–155). The author explains that by "scale," he means a thin skin formed during casting on the internal surfaces of ingot moulds. The causes of this phenomenon are discussed, and suggestions are made as to how it can be prevented.—L. J. L.

The Handling of Large Moulds and Flasks at Moulding, Assembling and Stripping Stations With Pneumatic Jacks. (*Fonderie*, 1955, Apr., 4483–4489).

Centrifugal Casting with a German Water-Cooled Machine. P. J. Ahern and J. F. Wallace. (*Foundry*, 1955, **83**, July, 130–135). A German centrifugal casting machine used for the production of gun barrels is described and its operation compared with that of a standard machine used at the Watertown Arsenal. The German machine uses a light steel, water-cooled mould lined with dry sand and directly supported by rollers. Experimental operation demonstrated that the equipment permitted greater latitude in pouring rate and spinning speed, and had a considerably higher production rate.—B. C. W.

Automation in Casting by the Vacuum Suction Method. N. A. Zhuikov. (*Liteĭnoe Proizvodstvo*, 1955, (3), 13–14). [In Russian].

The Application of Sub Sonic Vibrations During Solidification of Castings with Particular Reference to a Material for Gas Turbine Blades.—'H. R. Crown Max.' S. Hinchliff and J. W. Jones. (*College Aeronautics, Cranfield*, 1955, Apr., Report 89). The results of this study show that the process of vibration gives a smaller and more equi-axed crystal grain structure and that no expected improvement in mechanical properties are realized. The ultimate strength increase from 37·2 to 43·2 tons/in^2 and the elastic limit from 16 to 21·5. Details of the technique are given.—B. G. B.

Centrifugal Casting of Steel. (*Tecn. Indust.*, 1955, **23**, May, 579–580). [In Spanish]. Horizontal and vertical centrifugal casting techniques, the properties of centrifugally cast steel, and the advantages of the method for the manufacture of tubular products in special and alloy steels, are briefly discussed.—P. S.

Tyre Manufacture by the Centrifugal Casting Process. H. Dettela. (*Maschinenbau u. Wärmewirt.*, 1955, **10**, Apr.-May, 137–143).

Importance of the Croning Method Using Synthetic Sands and Resins in Modern Foundry Technique. F. Poelzguter. (*Fonderia Ital.*, 1955, **4**, May, 237–247). [In Italian]. The principles on which the Croning method is based are described. The development of the process is reviewed and the position relating to patents is clarified. The preparation of moulds and cores is described and details relating to moulding machines and their operation are given. Foundry sands, pouring techniques and the economics of the process are also discussed. (25 references).—M. D. J. B.

Shell Moulding. (*Indust. Gas*, 1955, **18**, Aug., 344–348). Details are given of the construction and use of the semi-automatic Shelmolda Duplas shell moulding machine.—D. H.

The Shell Mold Casting Process. (*Precision Met. Mold.*, 1955, **13**, Feb., 45–47, 50).

The Principles of Economic and Efficient Lay-Out of Sidings for Foundries. A. Rüppel. (*Met. u. Giesserei Techn.*, 1951, **1**, Feb., 54–60). The lay-out of sidings and engine sheds for all manner of foundries is discussed in some detail—L. J. L.

Compressed Air Economy in Foundries. J. Lempicki. (*Przeglad Odlewnictwa*, 1955, **5**, (4), 102–106). [In Polish].

The Importance of Thermal Economy in Foundries. T. Chabowski. (*Przeglad Odlewnictwa*, 1955, **5**, (3), 73–77). [In Polish].

Studies on the Fluidity of Molten Iron. III. Y. Koike. (*Tetsu to Hagane*, 1955, **41**, June, 587–592). [In Japanese]. Results are given for v of molten iron at 1250–1400° C, and

steel at 1525° C. v of iron is influenced by its state of oxidation and content of deoxidiser.—K. E. J.

Mould Reaction During the Casting of Non-Ferrous Metals. W. A. Baker. (*Gjuteriet*, 1955, **45**, June, 75–81). [In Swedish]. Mould reaction is defined as the reaction between the metal and the gaseous atmosphere in the mould. The occurrence of gas absorption in non-ferrous casting alloys is reviewed and examples are given of the influence of variables during casting on reaction extent. Possible mechanisms are discussed together with methods for suppressing the reaction. The influence of gas absorption on properties of castings is dealt with.—G. G. K.

Preventing Gas Absorption from the Mould When Casting Copper Alloys. T. Malmberg. (*Gjuteriet*, 1955, **45**, Apr., 45–50). [In Swedish]. Tests have shown that metal/mould reactions can be suppressed by using a mould material which will cause rapid solidification of the melt, or by spraying a protective metal coating on the mould wall. In both cases the density of the test bars employed increased, thereby indicating a reduction in gas absorption—G. G. K.

REHEATING FURNACES AND SOAKING PITS

Pusher-Type Billet Furnace at the Thionville Works of the Société Lorraine-Escaut. (*Centre Doc. Sidér., Circ. Inform. Tech.*, 1955, (8), 1563–1570). The new " Rust " type 2-zone billet reheating furnace with a capacity of 225 tons/8 h is described and illustrated. Operating characteristics are given in detail, including fuel consumption, combustion control, heat balance and heating time.—T. E. D.

Considerations on Pusher-Type Furnaces. Szczeniowski. (*Centre Doc. Sidér., Circ. Inform. Tech.*, 1955, (8), 1571–1591). Factors affecting the fuel efficiency are considered. Performances with various gas mixtures are compared, and the effect of preheating is assessed. Heat losses, gas flow, and problems of regulation and control are also discussed. Recommendations on dimensions, burner positions, method of charging, and choice of fuel are given.—T. E. D.

Some Aspects of the Reheating of Ingots of Special Steel by American Methods. M. Corbex and H. F. Spencer. (*Centre Doc. Sidér., Circ. Inform. Tech.*, 1955, (7), 1433–1446). The development of modern reheating furnaces is briefly discussed. Types of modern recuperator soaking pits and holding furnaces of American design are illustrated and discussed. Performance and heating times for various size ingots and slabs are given and discussed.—T. E. D.

Induction Heating of Ingots for a Steel Rolling Mill. (*Engineer*, 1955, **199**, May 13, 668–669; *Elect. Rev.*, 1955, **156**, May 20, 866–867; *Iron Coal Trades Rev.*, 1955, **199**, May 20, 1167–1168; *Sheet Metal Ind.*, 1955, **32**, Sept., 697–698). The first commercial plant for the induction heating of ingots for a steel rolling mill is described. Heating takes place by induction in two parallel tunnel furnaces through which the ingots are fed. Compared with the previous oil-fired furnaces the new installation is said to show 75% reduction in heating time, 30% increase in production and a saving of one dollar per ton in fuel or energy costs.—M. D. J. B.

Effect of Increased Output and Higher Quality Requirements on the Design, Operation and Control of Reheating Furnaces in Rolling Mills and Soaking Pits in Forging Shops. B. v. Sothen. (*Stahl. u. Eisen*, 1955, **75**, June 2, 709–718). A review based on published information mostly in *Stahl u. Eisen* on design, operation, and control of reheating furnaces and soaking pits. The author emphasizes the need of better furnaces for increased output and higher quality requirements; he deals with port, burner and roof design, shows flow-lines for various designs and discusses the instruments used in control of these furnaces.—T. G.

Electrical and Thermal Efficiency in Heating Cylindrical Bodies Inductively. H. Geisel. (*Elektrowärme-Technik*, 1955, **6**, May-June, 93–101). The principles of induction heating are discussed. Nomograms for the evaluation of optimum design of heaters for bodies in the shape of cylinders, i.e. steel bars, are given.—P. F.

Continuous Automatic Heating of Bars. (*Indust. Gas*, 1955, **18**, June, 277–284). Details are given of the construction and operation of a vertical gas-fired furnace for heating to 1000° C and continuous delivery of bars to a forming machine.—D. H.

HEAT-TREATMENT AND HEAT-TREATMENT FURNACES

The Tempering and Nitriding of Some 3% Chromium Steels. C. C. Hodgson and H. G. Baron. (*J. Iron Steel Inst.*, 1956, **182**, Mar., 256–265). [This issue].

Surface Protection of Titanium and Stainless Steel During Heat Treatment. N. Drover. (*Metal Progress*, 1955, **67**, June, 87–90). The use of extremely dry hydrogen for protecting the surface of stainless steel parts during hardening and annealing is described.—B. G. B.

Equilibrium Curves—Their Application to Furnace Atmospheres. N. K. Koebel. (*Steel Processing*, 1955, **41**, Apr., 247–253). The use of equilibrium curves in specific atmosphere heat-treating, where precision carbon control is desired, is discussed. The author reviews and compares theoretical equilibrium curves, calculated by means of the phase rule, with empirical curves recently derived by direct measurement methods in industrial furnaces. It is shown that theoretical curves do not agree with practical results due to inaccuracy of the equilibrium constant and the fact that true equilibrium never exists in industrial furnaces. The use of experimentally determined curves is therefore advocated. (7 references.)—P. M. C.

Control In Heat Treatment. J. G. Solomon. (*Australian Inst. Met.*; *Australasian Eng.*, 1954, Dec. 7, 54–61). The author gives an outline of the control methods normally used in heat-treatment. He covers the control of raw material, including acceptance tests for quality, the control of the treatment process (including temperature and automatic control) and of final inspection, notably hardness. (12 references).—P. M. C.

Service Failures Resulting From Defective Heat Treatment. L. E. Samuels and M. Hatherly. (*Australian Inst. Met.*; *Australasian Eng.*, 1955, Feb. 7, 61–64). A number of examples of service failures attributable to defective heat treatment are discussed, the examples being chosen to illustrate the commonest errors in heat treatment practice, viz. incorrect austenitizing temperature, carburization and decarburization during austenitizing, incorrect severity of quenching and incorrect tempering.—P. M. C.

A Brief Survey of Modern Heat Treating Equipment. J. E. Hyler. (*Steel Processing*, 1954, **40**, Dec., 793–795, 797). A brief review is given of equipment and techniques more generally used in modern heat treatment shops. Salt bath hardening, furnaces with integral quench tanks, protective atmospheres, combined brazing and hardening, and shaker-hearth furnaces with radiant heating are among the topics discussed.—P. M. C.

Gas Carburizing Atmospheres. L. G. W. Palethorpe. (*Wild Barfield Heat Treatment J.*, 1955, **5**, June, 10–15). The various separately generated and furnace generated atmospheres available for gaseous carburizing are described.

Experiences With Commercial Gas Carburizing Equipment. H. C. Thomas. (*Steel Processing*, 1955, **41**, Feb., 115–121). The author describes the operational procedures adopted by British Timken Ltd., for the carburizing of taper roller-bearing components made from low-carbon alloy steel (En 34). Three types of equipment are dealt with; pit furnaces for components above $2\frac{1}{2}$ in. dia., rotary drum furnaces for components below $2\frac{1}{2}$ in. dia., and a continuous gas carburizing scroll type furnace used on long production runs for a narrow range of component sizes with identical case depths. The choice of carburizing fluids for use in the batch type furnaces, and the various controls necessary throughout the processes are discussed.—P. M. C.

Gas Carburizing. L. G. W. Palethorpe. (*Wild Barfield Heat Treatment J.*, 1955, **5**, Mar., 8–12). The author discusses in some detail the theory of gaseous carburizing and details the advantages of the process.—D. H.

Thermo-Chemical Treatment of Steel in Molten Salts Containing Titanium. R. Skorski. (*Prace Instytutow Mech.*, 1954, **4**, (12), 23–35. [In Polish]. The physical properties of a molten mixture of sodium carbonate and titanium dioxide were investigated, and it was found that the strong carbonizing properties of these salts are due to the presence of titanium. It was also shown that steel thus treated is not only carbonized but obtains a fine structure. This is proportionate to increase in temperature of the bath. The structure of a white iron treated in this way becomes fine, the cementite being decomposed into spheroidal graphite and pearlite.—V. G.

Vacuum Heat Treating. R. L. Haff and A. M. Bounds. (*Steel*, 1955, **136**, Apr. 11, 108–111; Apr. 18, 108–110). The principles of vacuum heat treating are noted and typical industrial equipment described. Its use in treating the reactive metals is shown.—D. L. C. P.

Investigation of the Resistance to Wear of Sulphided High-Speed Steel Cutters. E. P. Nadeinskaya. (*Stanki i Instrument*, 1955, (2), 11–17). [In Russian]. The investigation described was carried out to check recent claims that sulphiding of tool steels increases their resistance to wear. Sulphiding was effected at 550–560° C for various times in a bath containing 17% NaCl, 25% $BaCl_2$ 38% $CaCl_2$ as the neutral portion, 13·2% FeS and 3·4% Na_2SO_4 as the active portion and 3·4% $K_4Fe(CN)_6$ as accelerant. Machining tests on a steel (0·45% C, 0·70% Mn, 0·31% Si, 0·08% Cr, 0·028% S, 0·024% P) showed that the wear-resistance of tools sulphided as above was greatly inferior to those which had been heat treated and cyanided.—S. K.

WELDING AND FLAME-CUTTING

Resistance Welding Control. C. J. Teece. (*Elect. Times*, 1955, **127**, Mar. 31, 523–527). The use of electronic equipment for switching and controlling the supply to a resistance welding machine is discussed.—L. D. H.

Presidential Address. H. G. Taylor. (*J. Trans. Soc. Eng.*, 1955, **46**, Jan.-Mar., 11–26). A survey of the development of resistance welding is given with special reference to the spot, seam, flash-butt, and projection welding processes. Modern techniques are dealt with briefly. (20 references).—L. E. W.

Automatic Hard Facing with Mild Steel Electrodes and Agglomerated Alloy Fluxes. J. S. McKeighan. (*Welding J.*, 1955, **34**, Apr., 301–308). The author discusses the mechanics of the hard facing process using a mild steel electrode and an alloy flux, equipment and its operation. A few field applications are described.—U. E.

Assessment of Spot-Weld Quality by Torsion Test. P. Joumat and J. E. Roberts. (*Brit. Welding J.*, 1955, **2**, May, 225–233). Experiments comparing a torsion test with other methods of assessing the quality of spot welds in low-alloy and plain carbon steel are described. It is shown that an inspection of the torsion failure gives sufficient information to decide whether the weld is satisfactory for service, without any need for mechanical measurements—U. E.

Inert-Gas Metal-Arc Cutting. R. S. Babcock. (*Welding J.*, 1955, **34**, Apr., 309–315). Laboratory tests have shown that the inert-gas metal-arc cutting process, using a consumable steel wire electrode with direct current reverse polarity, can be successfully used to sever and shape many non-ferrous metals, such as $\frac{1}{4}$ in. to $\frac{1}{2}$ in. aluminium plate, copper and nickel plate.—U. E.

Recent Developments on Contact Electrodes. D. L. Mathias. (*Welding J.*, 1955, **34**, Apr., 316–328). The characteristics of contact-electrodes are discussed, and compared with those of conventional types of electrodes. Deposition rate data are given and the effect of ageing on elongation is investigated.

Stainless Steel Fabrication Saves £350,000. T. A. Dickinson. (*Welding Metal Fab.*, 1955, **23**, May, 166–168). The inert gas-shielded arc welding equipment developed by Northrop Aircraft, Inc. of Hawthorne, California, is described.—U. E.

Application of Argonaut Welding to Carbon Steel. R. L. Fannon, and V. C. Herbert. (*Welding Metal Fab.*, 1955, **23**, May, 161–165). A report is given on research carried out by the British Oxygen Co. Ltd. and the United Steel Companies, Ltd. on the application of Argonaut welding to killed and semi-killed mild steel.—U. E.

Further Studies of the Flash Welding of Steels. E. F. Nippes, W. F. Savage, G. Grotke and S. M. Robelotto. (*Welding J.*, 1955, **34**, May, 223-s–240-s). Weld centerline rates of cooling at 1300°, 1000° and 900° F in flash welds in AISI 1020 and 4340 sheets, prepared under similar conditions, were measured and compared. It was found that the cooling rate is not influenced by the composition of the steel if the specimen geometry, platen acceleration, and final damping distance are maintained constant.—U. E.

Effects of Residual Stresses in Welded Structures. G. M. Boyd. (*Welding J.*, 1955, **34**, May, 216-s–222-s). The evidence for the effects of residual stresses arising from welding on the behaviour of mild steel structures is reviewed. It does not appear that residual stresses have any significant harmful effect, but that the detrimental effects are due to metallurgical changes, which in the presence of stresses lead to fracture.

Effect of Welding on Transition Temperature of Nickel Steel Plate. T. N. Armstrong and W. L. Warner. (*Welding J.*, 1955, **34**, May, 209-s–215-s). The properties of $3\frac{1}{2}$% nickel steel plate for welded pressure vessels used for temperatures as low as −150° F are discussed. The effect of heat treatment on fracture appearance of Charpy specimens was investigated. It was found that tempering or stress relieving slightly improves the Charpy impact properties of normalized $3\frac{1}{2}$% nickel steel plate at 70°, −100°, and −150° F. The transition temperature in the slow-notch bend or Virrizel test was slightly below −100° F for the as-welded normalized plate, but for the other three conditions the transition temperature was −150° F or below, the conditions being: quenched and tempered as welded; normalized, stress relieved after welding, quenched and tempered, stress relieved after welding; quenched and tempered stress relieved after welding.

Fatigue Aspects in Aircraft Welding Design. J. Koziarski. (*Welding J.*, 1955, **34**, May, 446–458). (77 References).

Iron-Powder Electrodes—A Progress Report. J. Hinkel. (*Welding J.*, 1955, **34**, May, 440–445). The use of iron powder electrodes is discussed and weld deposit properties are investigated. It was found that single vee butt welds made with a $\frac{3}{16}$ E—6027 electrode have excellent wash-in and cover pass appearance, and are less crack sensitive on medium carbon steels.—U. E.

Mechanization Applied to Oxygen Cutting. R. L. Deily. (*Welding J.*, 1955, **34**, May, 433–439). Automation in material handling and electrical controls for routine high-quantity production of precision work are described.—U. E.

Oscillographic Instruments in Spot Welding Quality-Control and Maintenance. G. Woodmancy. (*Welding J.*, 1955, **34**, May, 425–432). A cathode-ray oscilloscope and direct writing magnetic oscillograph are described.—U. E.

Hardenability Evaluation of Welding Electrodes. L. M. West. (*Welding J.*, 1955, **34**, May, 399–404). In the aircraft industry most weld assemblies involving hardenable steels are heat treated after welding. Testing procedures and test results for electrode deposits are reported.—U. E.

An Improved Method of Oxy-Fuel Combustion. E. H. Roper. (*Welding J.*, 1955, **34**, Apr., 337–344). The use of propane, methane or natural gas together with oxygen for welding, flame cleaning, flame hardening, oxygen cutting, etc., is discussed.—U. E.

Extension to Murex Laboratories. (*Welding Metal Fab.*, 1955, **23**, May, 183–186).

Rejuvenation of Roll Profiles by Welding. H. Charbonnier. (*Centre Doc. Sidér., Circ. Inform. Tech.*, 1955, **12**, (5), 1021–1030). A welding device is described in which the welding rod is automatically fed into the weld, and solidifies under an insulating layer which is fed as a powder on top of the weld. The current required for refacing rolls is 350–450A, depositing 3·5–4 kg/hour of metal.—T. E. D.

Long Welded Rails and Insulated Joints. C. W. Clarke. (*Rail. Steel Topics*, 1955, **3**, (1), 11–13).

Fifty Years of Welding in France. M. Lebrun. (*Mem. Soc. Ing. Civils France*, 1954, **107**, Apr.–June, 176–188).

Spot Welding From One Side—With a Portable Gun. C. A. McClean. (*Product Eng.*, 1955, **26**, Apr., 180–189). Using the inert-gas shielded tungsten-arc process, spot welding can be carried out with access to one side only of the work. The quality of the weld depends on several factors, such as the material thickness, surface condition, weld time, etc. Tests have shown that the welds obtained are equal in strength to those obtained by the conventional resistance process.

Welding of Plated Steels. L. F. Denaro and J. Hinde. (*Soudure Techn. Connexes*, 1955, **9**, Mar.-Apr., 63–80). The necessity for maintaining the continuity of the plating is emphasized, and methods are described, taking into account conditions of service, thickness of the plating, and its nature, namely nickel, Monel, Inconel or austenitic stainless steel. Applications and discussion are given.—T. E. D.

Spot Welding of Steels with a High Resistance. P. Joumat. (*Soudure Techn. Connexes*, 1955, **9**, Mar.-Apr., 97–100). A chart is given which enables spot-welding machines to be suitably adjusted.—T. E. D.

Cast Iron and Welding. S. M. Checa Casajus. (*Cien. Tecn. Solda.*, 1955, **5**, Jan.-Feb.). [In Spanish]. A brief review of the physical metallurgy of cast-iron is given, the problems involved in the successful arc-welding of cast-iron are discussed, and suitable types of electrodes are reviewed. It is recommended that a series of tests, as described, should be used

to decide the most suitable electrode. Joint preparation for a series of different welds is described.—P. S.

Joints Brazed With New Nickel Based Alloys Exhibit Good Properties. C. V. Foerster and R. C. Kopituk. (*Iron Age*, 1955, **175**, Feb. 24, 79–82). A new series of high-nickel brazing alloys is described and their applications explained. They are intended for brazing most stainless and heat resistant alloys, give joints with properties approaching those of the base metals and allow the introduction of modern production methods.—D. L. C. P.

Welding Electrode Research and Development. (*Metallurgia*, 1955, **51**, Mar., 145–148). A short account of a new research laboratory.—B. G. B.

Welding and Controlled Low Temperature. M. D. E. Robinson. (*Canad. Metals*, 1955, **18**, Feb., 42–44). The welding techniques used on 11 ft dia. steel penstock at Kemano, Canada are described in detail. Full particulars of the welding specification are given—B. G. B.

Surge Injection for A.C. Welding Transformers. L. H. Orton and J. C. Needham. (*Times Rev. Indust.*, 1955, **9**, Apr., 28–30). A method of surge injection is described which reduces the open circuit voltage for a.c. welding and eliminates or reduces any objectionable electrical effects such as radio interference.—B. G. B.

Fatigue of Spot Welds. G. Welter. (*Welding J.*, 1955, **34**, Mar., 153-s–156-s). The problem of fatigue resistance of untreated and hydrostatically treated spot welds is reviewed.

Coating Moisture Investigations of Austenitic Electrodes of the Modified 18-8 Type. K. P. Johannes, D. C. Smith and W. G. Rinehart. (*Weld. Res. Council, Weld. Res. Council Bull. Series*, 1955, Feb., (21)). Rates of moisture absorption by the coatings of three types and three brands of modified 18/8 austenitic electrodes were determined at temperatures of 75°, 85° and 100° F and relative humidities of approx. 50, 70, 82 and 96%. Temperature was found to be very effective in influencing the rate at which moisture is picked up by the coatings; the total amount absorbed at equilibrium was found to be determined by relative humidity. Moisture-loss determinations were also made.—U. E.

Pressure Vessel Welding—Production Up 93%, Costs Down 36%. L. C. Northard, Jun., (*Steel*, 1955, **136**, Jan. 10, 74–75). Inert gas metallic arc machine welding was found preferable to the hand arc method for making small mild or low alloy steel pressure vessels.—D. L. C. P.

MACHINING AND MACHINABILITY

French Contribution to the Analytical Study of the Cutting of Metals. F. Eugène. (*Mém. Soc. Ing. Civils France*, 1954, **107**, Apr.–June, 161–175).

The Scientific Study of Production Technique. O. Peters. (*Mém. Soc. Ing. Civils France*, 1954, **107**, Apr.-June, 89–112). The results of a number of studies concerning the mechanism of metal cutting and the formation of chips are discussed.

Ultrasonic Machining. I. G. Nishimura, Y. Jimbo and S. Shimakawa. (*Tokyo Univ., J. Faculty Eng.*, 1955, **24**, Mar., 65–100). Experiments on boring, cutting and finishing various hard minerals and alloys, using frequencies of 17–75 kc are reported. The preferred frequency range is 20–30 kc. Tool wear may be reduced by employing a tool larger than the hole to be bored. (37 references)—K. E. J.

Observe Six Rules when Making Tapered Shells. S. R. Cope. (*Machinist*, 1954, **98**, May 29, 950-952.)

Three Shell Jobs Typify Ironing Operations. S. R. Cope. (*Machinist*, 1954, **98**, Aug. 14, 1442–1444).

Necked Shells Require Careful Reduction. S. R. Cope. (*Machinist*, 1954, **98**, Sept. 4, 1564–1566).

Make Offsets in Dies or on Lathes. S. R. Cope. (*Machinist*, 1954, **98**, Oct. 8, 1806–1808).

Stampings often require Burred or Flanged Holes. S. R. Cope. (*Machinist*, 1954, **98**, Nov. 12, 2014–2017).

Three Kinds of Burrs Require Different Dies. S. R. Cope. (*Machinist*, 1955, **99**, Feb. 4, 210–213).

CLEANING AND PICKLING

Polishing of Metal by the Barrelling Process. L. Mable. (*Trans. Inst. Metal Finishing*, 1955, **32**, 289–298). Practical methods are described for barrel grinding and/or burnishing of malleable iron, steel, etc. Deburring and cleaning methods are given. The layout and organization of barrelling shops is also discussed.—A. D. H.

Instrumentation: Controlled Measurement Lowers Pickling Costs. D. H. Krouse. (*Iron Age*, 1955, **175**, Apr. 7, 132–134). The savings which can be obtained by accurate acid and water metering into pickling tanks are explained, and methods described.—D. L. C. P.

Blast Cleaning Provides Better Surface for Varied Mill Products. W. G. Patton. (*Iron Age*, 1955, **175**, Apr. 28, 108–110). The increasing use of abrasive blasting for surface cleaning of carbon and alloy steels is described. The problems of spent pickle acid disposal, and fume removal are thus avoided.—D. L. C. P.

Cleaning Room Production Control. H. A. Graber. (*Amer. Foundryman*, 1955, **27**, Mar., 34–37).

Wire Shot, the most recent Shot-Peening Medium. W. Gesell. (*Draht, German ed.*, 1955, **6**, Jan., 8–10). This is a survey of materials used in shot peening with the main emphasis on shot made from wire. Performance as function of wire quality and shot dimensions is described with particular reference to American practice.—J. G. W.

Automatic Pickling Installation for Wire Coils. A. Munz. (*Draht, German ed.*, 1955, **6**, Jan., 10–11). A mechanized pickling plant for the descaling of brass wire is described.

Electrolytic Salt Baths Descale Hot and Cold-Rolled Stainless Strip. T. J. Nolan and G. E. Rowan. (*Iron Age*, 1955, **175**, Mar. 3, 119–121). The continuous-line equipment installed in the new mill of Atlas Steels Ltd., Ont., for descaling both chrome-nickel and straight-chrome stainless steel strip is described. No scale breaking stage is necessary, for two electrolytic salt baths are used, the first oxidizing the scale and the second dissolving the oxides: a conventional acid pickle completes the process.—D. L. C. P.

Treatment of Spent Sulphuric Acid Pickling Liquors. A. M. Fradkin and E. B. Tooper. (*Indust. Eng. Chem.*, 1955, **47**, Jan., 87–90). A cation exchange method is described to treat spent sulphuric acid pickling liquors. The direct conversion of ferrous sulphate to sulphuric acid can be achieved. This acid can be recovered and re-used to reduce make-up requirements in the pickling tanks. The experiments carried out include a study of the more important operating variables and methods of regenerating the exchange resins. The recovery of other by-products from spent liquors is also discussed.—G. E. D.

Electrolytic Regeneration of Spent Pickling Solutions. H. C. Bramer and J. Coull. (*Indust. Eng. Chem.*, 1955, **47**, Jan., 67–70). Conventional electrolysis of highly acidic ferrous sulphate solutions results in preferential liberation of hydrogen at the cathode and low iron deposition. The process proposed makes use of ion exchange resins which are highly impermeable to water to inhibit the movement of H^+ ions towards the cathode. Iron deposition can therefore occur at high current efficiency in a low acid environment. Experimental work is described and important operating variables discussed.—G. E. D.

PROTECTIVE COATINGS

Anion Exchange Removal of Iron from Chloride Solutions. A. C. Reents and F. H. Kahler. (*Indust. Eng. Chem.*, 1955, **47**, Jan., 75–77). The industrial applications of anion exchange to remove iron from chloride solutions are discussed. In particular, the treatment of hydrochloric acid and aluminium chloride solution respectively were investigated. Quaternary type anion exchangers were used, the resin being regenerated with water. Data are given on exchange capacity, resin stability and process economics—G. E. D.

The Influence of Molten Metallic Coatings on the Mechanical Properties of Steels and Alloys. Ya. M. Potak and I. M. Shcheglyakov. (*Zhur. Tekhn. Fiz.*, 1955, **25**, (5), 897–907). [In Russian]. It was established that hardened steels and some other non-ferrous alloys are sensitive to the presence on their surface of molten metals, while austenitic steels, e.g. cadmium, lead, copper etc., are not sensitive to metallic coatings at temperatures slightly above their melting point. The causes of these differences in the behaviour and the mechanism of the action of metallic coatings on the mechanical properties of coated metals are discussed.—V. G.

A Study of Cracking in Chromium Deposits. H. Fry. (*Trans. Inst. Metal Finishing*, 1955, **32**, Advance Copy No. 2). The results of an exploratory experimental survey of cracking in chromium deposits from solutions of conventional compositions measurements of structure and stress of the coatings indicated that cracking may be associated with

structural abnormalities in the region of striations.—A. D. H.

Treatment of Chromic Acid Wastes. R. F. Ledford and J. C. Hesler. (*Indust. Eng. Chem.*, 1955, **47**, Jan., 83–86). Ion exchange methods can be used to recover chromium from the discharged rinse waters of plating baths. At the same time the rinse waters are rendered pure enough to comply with the anti-pollution laws.—G. E. D.

Diverse Properties Extend Engineering Uses of Nickel Plating. J. B. Mohler. (*Iron Age*, 1955, **175**, Mar. 10, 100–103). The various types of nickel plating baths and the properties of the deposit they produce are described. The hardness and ductility of the coating can be controlled. Electroless plating is useful for complicated shapes and internal areas.—D. L. C. P.

The Galvanization of Foundry Iron. W. Montgomery. (*Mét. Constr. Mécan.*, 1955, **87**, June, 535–540).

Galvanizing Plant Features Fume Gas Heating Controls. (*Western Metals*, 1953, **11**, May, 66–67). A new galvanizing plant at the Emsco Manufacturing Co., Los Angeles, is described. The steel and concrete tanks are heated by natural gas-fired immersion burners, and have a monthly capacity of 3000 tons.—P. M. C.

New Galvanizing Plant. Handling Techniques Eliminate Intermediate Storage Problem. (*Metal Ind.*, 1955, **86**, Jan. 14, 28–29). A brief description is given of a modern galvanizing production line at Quick Industries, Inc., Michigan, U.S.A. Six men handle an output of almost 3000 pounds per hour of 12 to 16 gauge mild steel sheet assemblies.—P. M. C.

The Simultaneous Electro-deposition of Two Metals with Special Reference to Tin Alloys. E. S. Hedges. (*Met. Ital.*, 1955, **47**, May, 197–199). [In Italian]. The theoretical aspects of electro-deposition of binary alloys are briefly reviewed. Practical applications are discussed and the progress made in the last few years in the field of electrolytic tin plating is described. (11 references).—M. D. J. B.

Research on the Electro-Chemical Behaviour of Tin. R. Piontelli, C. Serravalle and R. Ambrosetti. (*Met. Ital.*, 1955, **47**, May, 200–203). [In Italian]. A brief description is given of the experimental techniques used in studying the electrochemical behaviour of tin. A summary of results is given in which values of the current density between polycrystalline tin and various solutions are established. (8 references).

Hot Dip Aluminium Coating for Cast Iron. S. Ueda. (*Waseda Univ., Rep. Castings Res. Lab.*, 1955, (6), 38–41). Hot dip aluminizing of cast iron is more difficult than of steel, but if correct pre-treatment is carried out, coatings can be made without a flux. The influences of bath temp. and immersion time are elucidated. Thick alloy layers cannot be produced, even with prolonged low-temp. dipping. Al-coated cast iron has superior sulphurization-resistant properties, compared with conventional sulphurization-resistant irons, and good results were obtained in pilot-scale tests conducted in retorts used for the manufacture of CS_2.—K. E. J.

Aluminized Coatings. A. G. Thomson. (*Aircraft Eng.*, 1954, **26**, Aug., 266). A short review of the merits of hot-dip coating of aluminium and hot galvanizing for the protection of steel is presented. The results of work carried out by B.I.S.R.A. on hot-dip aluminium coatings are discussed.

Neoprene: Is It The Best All-Round Organic Coating? R. B. Seymour. (*Mat. Methods*, 1954, **40**, Dec., 93–95). The author proposes that neoprene is closest to the all-round "ideal" coating by comparing its performance with other well known materials from the point of view of adherence, ease of application, curing time, resistance to solvents, etc., strength and ductility, service life, and cost.—P. M. C.

The Assessment of Adhesion of Vitreous Enamels by Torsion. W. H. F. Tickle, M. K. Knowles and H. Crystal. (*Inst. Vitreous Enamellers Bull.*, 1955, **5**, May, 113–120). Apparatus is described for carrying out torsion tests on enamelled steel strips. Cracking of the coating was detected by a simple voltaic cell arrangement. The equipment is shown to give reproducible results and to give a sensitive index of the adhesion. It is considered to be complementary to impact testing.—A. D. H.

Cast Iron for Vitreous Enamelling. (*Inst. Vitreous Enamellers Bull.*, 1955, **5**, Mar., 89–98). The final report of a subcommittee set up "to study losses which are occurring in the wet process enamelling of cast iron and to report on their causes and to make recommendations for their elimination", contains the committee's findings. A bibliography and an atlas of defects are included.—A. D. H.

Mass Production Methods Spur Industrial Use of Refractory Ceramic Coatings. F. D. Shaw. (*Iron Age*, 1955, **175**, Apr. 21, 97–99). The properties, advantages and production of ceramic coated metals are briefly described.—D. L. C. P.

The Electron Microscope in the Study of Paints. N. D. P. Smith. (*Trans. Inst. Metal Finishing*, 1955, **32**, Advance Copy No. 12). The application of the electron microscope to problems of paint technology including pigment particle size and shape, the relation between surface contour and gloss and the nature of different forms of bloom are described.

POWDER METALLURGY

Powdered Iron Assembly Replaces Cast Iron. H. E. Mesnard. (*Precision Met. Mold.*, 1955, **13**, Feb., 36–37, 74). The advantages in economy accruing from the production of an acreage counter on a grain drill in powdered iron rather than cast iron are described.—D. H.

Can Sintered Steel be Evaluated from Machined Blanks. G. Zapf. (*Precision Met. Mold.*, 1955, **13**, Feb., 52–54, 70–72). It is economically advantageous to produce prototype components of sintered steel for test purposes by machining sintered blanks. The author discusses the advantages and possible disadvantages of this method with illustrations of typical case histories of components. Recommendations are given of suitable machining procedures.—D. H.

The Influence of Additions of Niobium and Tantalum Carbides on the Properties of Sintered Carbides. C. Agte and V. Dufek. (*Hutnické Listy*, 1955, **10**, (6), 322–329). [In Czech]. Effects of additions to carbides used for cutting tools are discussed on the basis of the authors' experiments and data given in the literature. Additions of 8% of TaC to WC–TiC–Co alloys increased their life as steel cutting tools by 100%.

Analysis of Continuity of One Phase in a Powder Mixture of Two Phases. F. Forscher. (*J. Franklin Inst.*, 1955, **259**, Feb., 107–114). A theoretical analysis of the continuity of one phase (powder A) in a mixture of two powders (A and B) is presented. It is assumed that powder A is distributed at random in powder B. For spherical particles of unequal size the continuity condition can be stated as the probability that the distance between the centres of any two particles of powder A equals the sum of their radii—B. G. B.

Studies of Grain Shape and Structure of Atomized Powders. E. Pelzel. (*Metall*, 1955, **9**, Feb., 81–88). An investigation of powders obtained by the atomization of melts of iron and copper powders is reported. In the case of iron reaction between the molten oxide and carbon monoxide causes hollow spheroids to be formed by atomization and this is not affected by additions of aluminium or silicon. In the case of copper the additions of other metals determine whether the grains are globular or otherwise depending on the behaviour of oxides on the surface of the metal jet.—J. G. W.

New Findings concerning Sieving Action. W. Batel. (*Z. V.d. I.*, 1955, **97**, May 1, 393–400). An investigation of the action of moisture in binding granular agglomerates, of the binding forces and of the sieving forces is reported, together with improvements of the design of moving sieves designed to improve sieving.—J. G. W.

The Rosin–Rammler Grain Size Distribution in Pulverized Powders. R. Brenner and A. Vidmajer. (*Metall*, 1955, **9**, May, 395–403). The Rosin–Rammler grain size distribution, which was arrived at empirically, is theoretically analysed, from which are indicated means of finding material constants and limitations of the distribution are discussed.—J. G. W.

The Nature of the Catalyst in the Decomposition of Carbon Monoxide in the presence of Iron. P. P. Das and B. Chatterjee (*Trans. Indian Inst. Met.*, 1953, **7**, 189–196). Iron powders obtained by the reduction of iron oxide with carbon monoxide and also by reduction of ferrous oxalate with hydrogen possess marked catalytic activity for the reaction $2\ CO = CO_2 + C$, the maximum precipitation of carbon taking place at 550° C. The activity of the catalyst decreased with time and approached zero when carbon monoxide was continuously passed over the powder. It is concluded that metallic iron and not iron carbide as present in white cast iron is the real catalyst in the reaction $2\ CO = CO_2 + C$.—D. H.

PROPERTIES AND TESTS

The Straining of Metals by Indentation including Work-Softening Effects. G. H. Williams and H. O'Neill. (*J. Iron Steel Inst.*, 1956, **182**, Mar., 266–273). [This issue].

A Study of the Impact Properties of Boron-treated Steel. S. J. Rosenberg and J. D. Grimsley. (*J. Iron Steel Inst.*, 1956, **182**, Mar., 278–292). [This issue].

Magnetic Sheet Iron with Orientated Crystals. (*Usine Nouvelle*, 1955, **11**, Apr. 7, 63–64).

A New Astatic Magnetometer for Measuring Ferromagnetic Sheet Test Pieces in Constant and Alternating Fields. K. Mellentin and H. Lange. (*Z. Metallkunde*, 1955, **46**, June, 450–456). An improved magnetometer is described, which is almost trouble-free in operation. A test sample is used in the form of thin ferromagnetic sheet, which for all practical purposes can be magnetized uniformly by an external field. Various applications of the instrument are described, with the relevant theory.—L. D. H.

On the Change of Electrical Resistance of High Carbon High Chromium Steels by Heat Treatment. T. Mutukura. (*Nippon Kinzoku Gakkai-Si*, 1952, **16**, Dec., 655–659). [In Japanese]. Resistivity curves are given for specimens slowly heated and cooled, and others air-cooled from different temp. The results are explained in terms of changes in structure. Details are also given for a steel containing 0·3% of W.—K. E. J.

Detection of Flaws in Metal with Ultrasonic Thickness Gauge. S. Hayashi, B. Kondo and H. Yamada. (*Bull. Eng. Res. Inst. Kyoto Univ.*, 1953, **4**, Sept., 54–57). [In Japanese].

Problems of the Manufacture and Ultrasonic Control of Heavy Upended Forgings. C. Roques, Ch. Dubois, and P. Bastien. (*Rev. Mét.*, 1955, **52**, May, 353–369). It is believed that inclusions can give rise to cracks previously initiated by forging. There is supporting evidence from the fact that in a transverse direction, there is a relative decrease in the metal's plasticity, which may lead to cold-working where the ingot is less pure. The danger of cracking is increased if the ratio of the diameter to height of the ingot is high. Ultrasonic examination clearly shows up such cracks but may not be able to distinguish them from cracks due to other causes.—G. E. D.

Ultrasonic Inspection of Semi-Finished Steel Products. H. Krautkrämer. (*Tekn. Tidskr.*, 1955, **85**, May 17, 457–461). [In Swedish]. The physical fundamentals of ultrasonic testing are outlined and details given of its application to the testing of forgings, tool steels, plate, wire and boiler tubes. Some developments in automatic testing are touched upon.—G. G. K.

Studies on the Accuracy Obtained in Ultrasonic Testing of Large Forgings. W. Türk, W. Knorr, and K. Barteld. (*Stahl u. Eisen*, 1955, **75**, May 19, 629–633). Operators of nine works, using their own equipment, tested a forging of 75 tons and an articulated spindle of 3·3 tons. Each team presented their findings independently of the others. The authors analysed these reports and examined the two forgings later by destructive testing. The findings of the different teams were almost identical, but sectioning of the suspected zones showed substantial differences from the ultrasonic testing results. The authors conclude that ultrasonic testing requires far more experience by the operator than any other testing method used in metallurgical practice.—T. G.

Ultrasonic Attenuation Measurements for Study of the Engineering Properties of Materials. R. Truell. (*Mech. Eng.*, 1955, **77**, July, 585–587). Brief details are given of the evaluation of ultrasonic energy losses in solids by the measurement of attenuation or propagation factor. The use of attenuation measurements in specific researches into the properties of some engineering materials is described.—D. H.

New Testing Methods Increase Foundry Efficiency. W. J. Stewart. (*Canad. Metals*, 1955, **18**, June, 40–42). A brief account of the use of non-destructive methods of testing castings is given.—B. G. B.

X-Ray Diffraction and Spectroscopic Analysis. Ursula M. Martius. (*Canad. Metals*, 1955, **18**, June, 22–26). A short account of the use of X-ray diffraction and X-ray spectroscopic techniques in metallurgy is presented.—B. G. B.

Studies of Non-Destructive Testing. (II) On the Dye Penetrant Inspection. S. Yoshida and K. Tamura. (*J. Mech. Lab.*, 1955, **9**, Mar., 47–50). A variety of dye penetrants for crack detection in bakelite, copper, brass, mild steel, and aluminium was tried. The technique proved equal to fluorescent penetrant inspection in sensitivity. Dyes employed included congo red, rhodamine and cellitone.—J. G. W.

An Application of X-Ray Diffraction Techniques. T. Geiger. (*Schweiz. Arch. Wiss. Techn.*, 1955, **21**, Feb., 50–55). The X-ray diffraction method was used to examine creep in two nickel-chrome steels. Relatively rapid creep involved deformation by slip but slow creep in one of the steels produced cellular break-up previously demonstrated by Wood for aluminium.—J. R. P.

Non-Destructive Testing. F. Rohner. (*Schweiz. Arch. Wiss. Techn.*, 1954, **20**, Nov., 371–373). A review is made of the progress and application of non-destructive testing since 1948, with 18 references. Under radiography, the γ-ray source isotopes Cobalt 60, Iridium 192, and Caesium 137 are discussed as well as the β-ray source Thulium 170. The application of magnetic powder, induction and field strength detectors is described for the location of flaws by the detection of discontinuities in an applied magnetic field. Electric-induction and ultrasonic testing techniques are also reviewed.

Measurement of Thermal Conductivity According to Diesselhorst. O. Rüdiger and H. D. Dietze. (*Tech. Mitt. Krupp*, 1955, **13**, July, 56–61). The theory and method of Kohlrausch and Diesselhorst, by which thermal conductivity can be determined by increase of resistance in passing an electric current is discussed, and described in detail.—L. D. H.

Studies on the Hot-Working Strength of Steels. II. K. Inouye. (*Tetsu to Hagane*, 1955, **41**, June, 593–601). [In Japanese). Using a high-temp., high-speed tensile testing machine (*ibid.*, 1955, **41**, May, 506–515), tests were made at 830°, 1030° and 1230° C, at various speeds, on 15 types of steel to be used for seamless tubes. Nominal stress/strain and true-stress/natural-strain curves were obtained, and values of deforming stress (at 20% stress) are given. An equation is developed between deformation stress, strain, deforming speed and temp.—K. E. J.

Study on Heat-Resisting Steel. VIII. E. Asano. (*Tetsu to Hagane*, 1955, **41**, June, 607–613). [In Japanese]. The solution of pptd. particles was studied in Timken 16-25-6, by varying the C, Mo and N_2 contents, and, before solution treatment, fully or almost completing pptn. in two ways: (*a*) heating at >950°, 900° or 850° C for several h and then holding at 700° C for >100 h, and (*b*) ageing at 800° C for 100–150 h. Three temperatures were used for solution treatment in each case. The results of hardness and microscopical examinations are given.—K. E. J.

Effect of Microstructure on the Elevated-Temperature Strength of Alloy Steel. J. J. Heger, J. M. Hodge and P. W. Marshall. (*Proc. Amer. Soc. Testing Mat.*, 1954, **54**, 1003–1027). The tensile properties at room temp. and at 1000°, 1100° and 1200° F, the ductility-transition temperature, and the creep-rupture strengths at 1000°, 1100° and 1200° F of a 2·5% Cr, 0·5% Mo, steel have been studied as a function of microstructure. Coarse pearlite and ferrite, fine pearlite and ferrite, upper bainite, lower bainite, and martensite were studied.—B. G. B.

Study on Heat-Resisting Steels. VII. E. Asano. (*Tetsu to Hagane*, 1955, **41**, May, 524–531). [In Japanese]. Hot-forged Timken 16-25-6 alloys, containing varying N_2 contents (including zero), were solution-treated and aged at different temp. for varying times. In general, solution treatment at higher temperature for long periods produced lower hardnesses before and after ageing, satisfactory removal of the influence of forging, larger grain size, and a more representative structure. Hardnesses equivalent to those in specimens initially given a shorter, lower-temperature solution treatment could, however, be produced by a suitable choice of ageing conditions. K. E. J.

Studies on the Hot-Working Strength of Steel. I. The High Temperature-High Speed Tensile Testing Machine. K. Inouye. (*Tetsu to Hagane*, 1955, **41**, May, 506–515). [In Japanese]. A description is given of a stress-strain testing machine, designed to produce results comparable with those obtaining during hot-rolling, forging, extrusion, etc., and the experimental technique is described. (11 references). K. E. J.

TOR-Stahl 40 at Fire Temperatures. S. Soretz. (*Schweiz. Arch. Wiss. Techn.*, 1955, **21**, Jan., 3–7). Tests are described in which TOR-Stahl 40 and mild steel were heated under tensile stress. The loads did not exceed permissible stresses at normal temperatures and the heating by electric current produced temperatures to be expected under conditions of fire with reinforced concrete structures. Neither TOR-Stahl 40 nor mild steel withstood temperatures above 500° C unless unstressed but when this temperature was not exceeded the TOR-Stahl 40 showed no alteration of the proof stress or ultimate strength when tested after cooling.—J. R. P.

How Important is Hydrogen Embrittlement? N. J. Grant and J. L. Lunsford. (*Iron Age*, 1955, **175**, June 2, 92–94). Experiments showing the effect of hydrogen absorption on embrittlement of mild steel are described. In cold-worked steel the ductility is permanently affected and the normal fracture pattern altered, while the hardness is not increased. The effect on hot-rolled material is not so severe.—D. L. C. P.

The Solution of Gases in Liquid and Solid Metals. A. E. Jenkins. (*Australian Inst. Met.: Australasian Eng.*, 1955, Feb. 7, 51–60). Recent investigations concerning the solubility of gases in metals are summarized. The fundamental laws of solution are discussed and it is shown that an analysis using methods of statistical mechanics can lead to a more complete understanding of the solution process. Attention is drawn to the difficulties associated with experimental techniques. (65 references).—P. M. C.

Effects of Hydrogen at High Pressures on the Mechanical Properties of Metals. H. C. Van Ness and B. F. Dodge. (*Chem. Eng. Prog.*, 1955, **51**, June, 266–271). Experiments on hydrogen embrittlement at low temperatures and hydrogen attack at high temperatures were performed to ascertain which metals could be used safely with high pressure hydrogen (up to 4000 atm. at room temp. and 2000 atm. at 425°–475° C). Preliminary results indicate that the most suitable metal is austenitic stainless steel.—T. E. D.

Effect of Arsenic, Copper, Tin and Molybdenum on the Impact Properties of Steels. H. Sawamura and T. Mori. (*Mem. Fac. Eng. Kyoto Univ.*, 1954, **16**, Oct., 205–227). Results are reported from investigations of the effects of As on the workability of C steels and of As, Cu, Sn and Mo on the workability of dead-soft steel. 47 specimens were melted, forged and annealed, and Charpy tests were made at temp. up to 900° C; the structures, ferritic and austenitic grain sizes, and hardnesses were also investigated. (18 references).—K. E. J.

Testing and Inspection. (*Iron Age*, 1955, **175**, June, K2–K12).

The Cracking of Layers of Brittle Material by Differential Strains. D. J. Millard. (*Brit. J. Appl. Phys.*, 1955, **6**, Apr., 124–127). The subject has been investigated, both theoretically and experimentally, to determine the factors which govern the pattern of the cracks produced by a given strain pattern.—E. E. W.

Diffusion and Oxidation of Solid Metals. C. E. Birchenall. (*Indust. Eng. Chem.*, 1955, **47**, Mar., 604–613). A critical review of recent work on the diffusion and oxidation of ferrous and non-ferrous solid metals and alloys. The diffusion rates of cobalt, chromium and wolfram in commercial iron and steel are discussed as well as the oxidation of pure iron, commercial iron and low-alloyed steels under various conditions of temperature and pressure. (115 references).—G. E. D.

Relaxation Effects in Solid Solutions Arising from Changes in Local Order. II. Theory of the Relaxation Strength. A. D. Leclaire and W. M. Lomer. (*Acta Met.*, 1954, **2**, Nov., 731–742). An expression is derived for the strength of relaxation arising from the stress-induced redistribution of atoms in substitutional solid solutions which is in satisfactory agreement with measured values. It is critically compared with an earlier theory due to Zener.—A. D. H.

The Dynamic Straining of Metals having Definite Yield Points. D. B. C. Taylor. (*J. Mech. Phys. Solids*, 1954, **3**, Oct., 38-46). Strain rate during non-uniform yielding of metals under repeated dynamic loading is analysed, with the result that the observed time/yield stress relationship is a 'delay period' effect, so that the dynamic yield criterion appears of the form 'stress-time integral constant', possibly with the addition of a term dependent exclusively on stress.

A Determination of Plastic Stress-Strain Relations. B. B. Hundy and A. P. Green. (*J. Mech. Phys. Solids*, 1954, **3**, Oct., 16–21). By pulling notched strips past their plastic yield point, according to a method suggested by Hill, the plastic potentials of copper, zinc and stainless steel were determined and found to be in agreement with the v. Mises potential.

A Theory of the Plastic Yielding due to Bending of Cantilevers and Fixed-Ended Beams. A. P. Green. (*J. Mech. Phys. Solids*, 1954, **3**, Oct., 1-15; 1955, **3**, Jan., 143–155). Plane strain and plane stress solutions are obtained for the plastic yielding of horizontal uniformly tapering cantilevers of rectangular section carrying a vertical end load. The effects of weak end support, and of axial loading in addition to vertical loading are analysed for uniform cantilevers in plane strain and plane stress. The solutions are also applied to centrally or uniformly loaded beams, either fixed at both ends or freely supported, and to beams of I-section bent about an axis perpendicular to the web. Satisfactory agreement with experiment is quoted.

A Method for the Determination of Poisson's Ratio by the Direct Measurement of Lateral Contraction and Linear Extension. K. Claus. (*Z. Metallkunde*, 1955, **46**, Aug., 589–592). Lateral contraction is determined by inductivity measurements, and linear extension is measured simultaneously. Poisson's ratio is determined with an accuracy of about 5% for a lateral contraction $2 \cdot 10^{-5}$ mm/mm.—L. D. H.

A Remotely Operated Extensometer. R. G. Berggren and J. C. Wilson. (*Amer. Soc. Test. Mat.*, Preprint, 1955, No. 90). A sensitive extensometer for use in remotely controlled tension testing of radioactive metallic specimens is described.—B. G. B.

Steep Decline in the Notch Toughness Curve and the Fracture Face of Notch Impact Test Pieces of Annealed Steel Plate. H. Kornfeld. (*Stahl u. Eisen*, 1955, **75**, Mar. 10, 265–278). Izod and Charpy V-notch impact tests were performed on annealed plates of killed and rimming common and fine-grain open-hearth steel and of basic Bessemer fine-grain steel over a wide temperature range (+ 80 to −130° C). The formulae for relative mean impact toughness ($\bar{K}_T$ (δ)) and probability (W_m (δ)) for the occurrence of composite fracture, i.e. partly brittle and partly ductile, are modified by substituting the steep decline by a straight line in the $\bar{K}_T$ (δ) curve. The author found a clear relationship between energy absorbed by the test bar and the fracture face in the impact tests and the steep decline of the notch impact curves as a function of temperature. He concludes that an appraisal of a given steel will be correct only if the required testing temperature is not lower than the range at which composite fracture occurs. T. G.

The Yield of Mild Steel Under Impact Loading. J. D. Campbell. (*J. Mech. Phys. Solids*, 1954, **3**, Oct., 54–62). Results of impact tests on cylindrical mild steel specimens are quoted. The derived dynamic stress/strain curve indicates that the upper and lower yield stresses are about double those obtaining under static conditions. The results are comparable with those of other investigators obtained by markedly different methods of testing.—J. G. W.

Residual Stresses in Heavy-Wall Cylinders. J. H. Faupel. (*J. Franklin Inst.*, 1955, **259**, May, 405–419). The amount of residual stress resulting from the autofrettage of steel cylinders appears to be related to the hardness and/or structure of the steel employed. Results presented show that an unalloyed steel cylinder autofrettaged to raise its elastic breakdown pressure significantly did not indicate residual stresses sufficient to account for the measured increase in elastic breakdown pressure. Residual stresses formed by thermal shock are sufficient to account for the change in elastic breakdown pressure measured in a pressure test regardless of the hardness of the material.—B. G. B.

Theoretical Analysis of Stress due to Moving Heat Source. M. Watanabe and K. Satoh. (*Osaka Univ., Techn. Rep.*, 1954, **4**, Oct., 365–379). The problem of plane thermal stress due to a moving heat source is examined in general, within the confines of elastic stresses, and the analysis is applied to the heat source with straight-line motion at constant speed and applied to an infinitely large plate.—J. G. W.

Effect of Various Stress-Relieving Treatments on the Behaviour of Welded and Notched Mild-Steel Specimens Under Impact Loading. G. Coates. (*Brit. Welding J.*, 1955, **2**, June, 266–275). The paper describes tup impact tests on notched steel bars into which residual welding stresses have been introduced. A comparison is made between the effectiveness of various stress-relieving treatments in improving impact performance. Heat treatment at 650° C is shown to have a greater influence than other methods of stress-relief.—U.E.

Investigation of the Effect of Heating on Cold Working and Residual Stresses. K. Wellinger and E. Keil. (*Schweiz. Arch. Wiss. Techn.*, 1954, **20**, Aug., 264–268). Tensile creep tests and torsional relexation tests were carried out on specimens cut from boiler plate, both as delivered and after cold-stretching 20%. There was no marked difference in tensile creep rates, but up to 450° C the permanent deformation was less for the cold-stretched material. In the relaxation tests, heating to 350° C showed that yielding first occurred at an appreciably greater load for the cold-stretched material. Cooling and reheating produced no further decrease in the torque supported.—H. R. M.

Laboratory Installation for Determining the Tendency of a Steel to Hot Cracking. V. M. Shpeizman. (*Liteinoe Proizvodstvo*, 1953, (7), 12–13). [In Russian]. A special testing machine is described for evaluating the tendency to hot cracking of steel specimens. During a test the temperature of the specimen, the stress to which it is subjected and the free linear contraction during cooling from the crystallization temperature to room temperature are evaluated. The method is that two similar specimens are cast simultaneously and are then allowed to cool. The contraction and temperature of one, which is allowed to contract freely, is recorded. The contraction of the other is prevented by an applied load which is measured and recorded. A series of tests is also described in which tendency to hot cracking was evaluated from the behaviour of frame-shaped specimens. Results for several alloy steels are presented.—S. K.

Black Spot Rail Defects on the Rhodesia Railways. B. M. E. O'Mahoney. (*Rail. Steel Topics*, 1955, **3**, (1), 14–20). In 1952 it was noticed that black spots at which were the sites of internal cracks had appeared on 3% of the rails in the Rhodesia Railway system. The experience of other railways, statistical and metallurgical investigation were of little direct use in solving the problem which appeared to be connected with high pressure causing distortion on the running edge of the rail head. Although a reduction in locomotive running speed was a palliative, the problem was never solved but the incidence of ' black spot ' declined.—A. D. H.

On the Relation between Structure and Strength. I. On the Statical Failure of Steel and Cast Iron. II. On the Fatigue Failure of Steel and Cast Iron. F. Matsumoto. (*Nippon Kinzoku Gakkai-Si*, 1951, **B. 15**, Sept., 395–402; 402–407). [In Japanese]. Cases of ductile and brittle failure are considered; formulae are derived for the yield point and proportional limit of steel, and for the tensile strength of cast iron, in terms of the amounts of the structural phases present and the properties of each constituent.

Similarly, an expression is given for fatigue strength in terms of ' separating strength ', and an approximation is derived for steel in terms of % C and the properties of the phases present. The latter relationship is applicable to carburized steel. (34 references).—K. E. J.

Improvement in Behaviour of Parts under Repeated Strains by Induction Hardening. (*Aciers Fins Spéc. Franç.*, 1955, Mar., 32–36). The presence of residual stresses in a hardened machine part is discussed, and reference is made to the origin of fatigue fractures. Induction hardening under controlled conditions has been shown to lead to a considerable increase in the limit of fatigue. A number of precautions to be taken essential to satisfactory results are also discussed.—E. A. C.

Elastic Constants and Coefficients of Thermal Expansion of Piping Materials Proposed for 1954 Code for Pressure Piping. R. Michel. (*Trans. Amer. Soc. Mech. Eng.*, 1955, **77**, Feb., 151–159). Values are given for the elastic constants and coefficients of thermal expansion for various piping materials at temperatures where applicable up to 1200° F. Some details are given of the test methods used and the accuracy of the results.—D. H.

Piping-Flexibility Analysis. A. R. C. Markl. (*Trans. Amer. Soc. Mech. Eng.*, 1955, **77**, Feb., 127–149). The author discusses the problem of plastic flow and fatigue in piping systems under thermal expansion as opposed to the ordinary steady state room temperature problem.—D. H.

Carbon-Molybdenum Steel Steam Pipe after 100,000 hours of Service. R. J. Sinnott, I. A. Rohrig, J. W. Freeman and A. I. Rush. (*Trans. Amer. Soc. Mech. Eng.*, 1955, **77**, July, 779–787). Carbon-molybdenum steel pipe carefully measured for service creep during 100,000 h of operation at 900° F was subjected to laboratory examination after removal from service. Remarkable correlation was observed between calculated service creep rates and those established by subsequent laboratory creep testing.—D. H.

Safe Stress Range for Deformation Due to Fatigue. M. Kawamoto and K. Nishioka. (*Trans. Amer. Soc. Mech. Eng.*, 1955, **77**, July, 631–634). The safe stress range of materials for fatigue is determined from both fatigue destruction and fatigue deformation. The authors believe that the safe stress range usually considered for fatigue deformation is incorrect for materials with no clear yield point and suggest other constants based on their experimental results.—D. H.

Military Tank Torsion Bars Shot Peened for Fatigue Resistance. E. Brooker. (*Western Metals*, 1953, **11**, Oct., 61–63). Steps in the production of torsion bar springs up to 96 in. long × 2 in. dia. are described and illustrated. The springs are made from hot rolled chrome-nickel-molybdenum steel, and are shot peened to improve fatigue resistance.—P. M. C.

The Dimensions of Machine Parts Submitted to Fatigue. Experimental Contribution to the Study of the Size Effect and the Effect of Notches. C. Massonnet. (*Rev. Univ. Min.*, 1955. 9th series, **11**, June, 203–222). The design of machine parts to ensure satisfactory fatigue resistance in service is first considered. Details are given of work on fatigue carried out at the University of Liège on carbon steels. The use of electrolytic polishing of fatigue specimens is explained. The tension gradient in specimens has a considerable influence on fatigue resistance. The influence of notches in specimens with a fine grain size is in agreement with the Peterson law. A large grain size reduces the harmful effect of notches on fatigue resistance.—B. G. B.

Studies on the Fatigue of the Metals. V, VI. Studies on the Fatigue by the Application of Magneto Striction. S. Tuda. (*Nippon Kinzoku Gakkai-Si*, 1951, **B. 15**, Oct., 457–460; 460–464). [In Japanese]. Results are presented of investigations in which the progress of fatigue in worked and non-worked metals is followed by the inverse Wiedemann effect. In C steel, pure Fe and Ni specimens it was found that changes in the induced current were similar whether the load was above, equal to, or less than the fatigue limit. Residual stress was very small in comparison with that produced by cutting the specimen on a lathe.—K. E. J.

The Influence of Carbon Content on the Hardness of Carbon Steels at High Temperatures. M. G. Lozinskii and S. G. Fedotov. (*Zhur. Tekhn. Fiz.*, 1954, **24**, (9), 1609–1612). [In Russian]. Experimental determination of the influence of the carbon content in ordinary carbon steels on their hardness in the temperature region room temperature–860° C is described.—V. G.

Experiments in Standardising the Scleroscope Hardness Test. H. Schmitz and W. Schlüter. (*Stahl u. Eisen*, 1955, **75**, Apr. 7, 411–416). Tests with 31 Scleroscope hardness testers showed that the results scatter up to 20° Shore for the same test-piece, whereas the individual hardness testers showed a scatter of 5° Shore. A standardization of the test indications by shifting the scales could not be obtained as there is no clear relationship between Rockwell C hardness and Scleroscope hardness. Tests with five new instruments of one make were also unsatisfactory. The authors suggest the introduction of standard test pieces of various hardness which should be used for the determination of correction factors of each individual hardness tester.—T. G.

Current Knowledge of the Ageing of Mild Steels. W. Wepner. (*Arch. Eisenhüttenwesen*, 1955, **26**, Feb., 71–98). The author reviews recent published information on the ageing of mild steels under the headings: Results of Experiments; Diffusion and Solubility of Companion Elements in Iron; Theories on Ageing Phenomena with emphasis on Dislocation Theories. (123 references).—T. G.

A Study of the Cold Brittleness of High-Purity Ferrite. M. Okada, S. Nakanashi and Z. Baba. (*Osaka Univ., Techn. Rep.*, 1954, **4**, Oct., 347–355). The influence of testing temperature and loading velocity on the cold brittleness of high purity ferrite was studied and the mechanisms of fracture identified.—J. G. W.

The Problem of Brittle Fracture in Steel. W. Felix and T. Geiger. (*Schweiz. Arch. Wiss. Techn.*, 1955, **21**, Feb., 33-49). The authors describe investigations of the process of fracture in the notch impact test. X-ray diffraction photographs showed that at low notch impact energies, insufficient for fracture, some of the crystals in the surface of the notch had sustained brittle fracture while others were intact. With increase in impact energy further crystals were deformed until a microscopic crack appeared. This mixed microscopic fracture condition was also found in the fractured surface of an H girder destroyed by brittle fracture in service and it was shown that the undeformed crystals especially those cold worked in the fracture tended to be orientated in the same plane.—J. R. P.

Welding and the Materials Problem. The Brittle Fracture of Steel. A. Matting. (*Schweissen u. Schneiden*, 1955, **7**, June, 270–274). The problem of brittle fracture of steel is discussed. It is shown that the notch impact test is the most suitable test method for determining the tendency to brittle failure. A variation in testing procedure is proposed where a

defined load is super-imposed and the steel is more easily differentiated into more or less sensitive steels.—U. E.

Brittle Fractures and Strength Along the Thickness of Steels. W. Soete. (*Rev. Mét.*, 1954, **51**, Dec., 813–823). In some tests on notched test-pieces, fissures appear in a direction parallel to the principal stress, leading to a complete alteration in the state of strain, which ceases to be tri-axial. A transition temperature above which fissures parallel to the rolling direction will appear, can be defined by comparing the ratio of the strength in the direction of the thickness to that in the longitudinal or transverse directions, with the ratio of the corresponding strains.—G. E. D.

The Fracture Mechanism of Steels Intended for Welding. C. Schaub. (*Jernkontorets Ann.*, 1955, **139**, (3), 154–173). [In Swedish]. A summary of previous work on brittle fracture is followed by an account of the determination of the critical nominal tensile stress for a Si Al-killed acid Bessemer steel, V-notched test plates being subjected to pulsating tensile stresses until fatigue cracking and brittle fracture ensued. Results showed that compared with the Charpy SIS test the Charpy V test permitted a better assessment of the steel's tendencies towards brittle fracture where thin 12 mm plates were concerned, but that fracture in thicker plates was facilitated—regardless of the internal structure of the material—by the geometric effect.—G. G. K.

Analysis of the Effect of Various Factors on Metal Transfer and Wear between Specimen Pairs of Same Metal and Same Shape. I. The Basic Scheme of Formation of Metal Transfer and Wear. I—Ming Feng. (*J. Appl. Phys.*, 1955, **26**, Jan., 24–27). The author describes the basic terms used in formulating metal transfer and wear. These are potential amount of metal transfer, and the mechanical, thermal and adhesion factors of the chance of metal transfer.—E. E. W.

Magnetic Balance for Measuring Susceptibility. H. Beisswenger and E. Wachtel. (*Z. Metallkunde*, 1955, **46**, July, 504–507). The construction and principle is described of a magnetic balance for para-, dia-, and ferromagnetic measurements on metals. Using 1·5 g of sample a specific susceptibility of 1×10^{-7} can be measured to an accuracy of 1%. Ferromagnetic constituents can be detected down to an order of magnitude of 4×10^{-8} cm³.—L. D. H.

Eddy-Current Phenomena in Ferro-Magnetic Materials. H. M. McConnell. (*Trans. Amer. Inst. Elect. Eng.*, 1954, **73**, Part I, 226–235). A theory is developed to demonstrate a logical transmission from the linear to the non-linear theories of eddy currents in solid media, the method being applied to typical problems.—L. D. H.

Domains of Reverse Magnetization in Ferromagnetic Salts. T. G. Nilan and W. S. Paxton. (*Phys. Rev.*, 1955, **97**, Feb.1, 834–835). A study of the magnetic powder patterns in polycrystalline grain-oriented silicon steels in which the authors state that Goodenough's calculations of the nucleation energies at the grain boundaries and at crystal surfaces are not applicable to the types of material and sample geometry usually encountered in power transformer applications.—E. E. W.

Magnetic Properties of FeTe. E. Uchida and H. Kondoh. (*J. Phys. Soc. Japan*, 1955, **10**, May, 357–361). The susceptibility-temperature relation of FeTe was measured over the range between liquid air and 1150° C. Antiferromagnetic behaviour was observed below 500° C, there was evidence of a feeble constant paramagnetism above 600° C, with an anomaly at the melting point, 815° C. A tentative explanation of these properties in terms of the electronic configuration of Fe is given.—J. G. W.

A Contribution to the Question as to Whether Magnetic Fields Affect the Hardening of Carbon Steel. F. Erdmann-Jesnitzer. (*Met. u. Giesserei Techn.*, 1951, **1**, Feb., 38–39). Attempts have been made to measure the effect of a magnetic field on steel samples. The author concludes that the properties of carbon steel cannot be improved by magnetization associated with hardening.—L. J. L.

Magnetic Materials with Rectangular Hysteresis Loops. B. G. Parkin. (*P.O. Elec. Eng. J.*, 1955, **48**, Apr., 1–6). The theory explaining the formation of rectangular hysteresis loops is given, applications described and the characteristics of 10 materials tabulated. (35 references).—L. D. H.

A Method for the Measurement of Constant Magnetic Fields and Constant Field Differences and its Application to Metal Research and Technique. F. Förster. (*Z. Metallkunde*, 1955, **46**, May, 358–370). The theoretical basis is given for the use of Förster probe coils for the measurement of constant field strengths and field differences. A number of applications are described, including classification of steel parts by the measurement of residual fields, measurement of wall thickness, detection of iron in non-magnetic material and examination for cracks.—L. D. H.

Effect of Low Temperature on the Stability of Permanent Magnets. A. G. Clegg. (*Brit. J. Appl. Phys.*, 1955, **6**, Apr., 120–123). The results are given of magnetometer measurements, at temperatures between + 60° and − 60° C, on five types of magnet materials having a range of dimension ratios. The influence of tempering is discussed.—E. E. W.

Thermoelectric Power and Electrical Resistivity of Dilute Alloys of Silicon in Copper, Nickel and Iron. C. A. Domenicali and F. A. Otter. (*J. Appl. Phys.*, 1955, **26**, Apr., 377–380). Data are given, by means of seven graphs, for temperatures between −195° and + 500° C.—E. E. W.

Testing of Materials by Ultrasonic Methods. J. Ors Martinez. (*Inst. Hierro Acero*, 1955, **8**, Apr.–June, 184–201). [In Spanish]. The method is described, and the types of apparatus commercially available are reviewed. An illustrated section deals with the classification of echoes and gives practical examples. The correct examination technique for a range of products is given in tabular form. (24 references).—P. S.

The Use of Sonic and Ultrasonic Methods in Metallurgy. H. Mathiske. (*Metallurgie*, 1955, **5**, May, 165–169). The author discusses sonic and ultrasonic methods to improve the quality of melts, and ultrasonic methods of waste gas purification. The equipment manufactured by the Ultrasonic Corporation is briefly described.—L. J. L.

Signalling, Counting or Recording the Defects of Metallurgical Products by Means of Automatic Ultrasonic Echosounding. L. Beaujard, V. Husarek and A. Vasset. (*Rev. Mét.*, 1955, **52**, Mar., 240–248). An account is given of the various devices used to select echoes with ultrasonic sounding apparatus and some examples given of its metallurgical applications.

Ultrasonics Makes Itself Heard. P. K. Bloch. (*Steel*, 1955, **136**, May 16, 118–119). The use of ultrasonics for differentiating between nodular and gray iron, and for thickness measurement is described. Thicknesses may be determined to within 0·1—1·0%.—D. L. C. P.

The Determination of the Elastic Constants of Metals by an Ultrasonic Method. E. G. Ramachandran and N. Srinivasan. (*Trans. Indian Inst. Met.*, 1953, **7**, 173–188). Young's Modulus and the Rigidity Modulus of mild steel have been determined as $18{\cdot}1 \times 10^{11}$ dynes/cm² and $6{\cdot}60 \times 10^{11}$ dynes/cm² respectively. A resonant vibration method was used, the vibration being produced by piezo-electric wedges driven by a valve oscillator.—D. H.

Developments in Non-Destructive Testing. R. H. Warring. (*Mech. World*, 1955, **135**, Feb., 87–89). A brief history of ultrasonic flow detection is given and the latest developments described. The use of eddy currents for the determination of flows in conducting materials is mentioned.—D. H.

The Scope of Gamma-Radiography. L. Mullins. (*J. Brit. Inst. Radio Eng.*, 1955, **15**, Mar., 131–141). The relative merits and scope of γ-rays and X-rays are discussed, and the applications suitable to each technique are indicated.

A High-Temperature, X-Ray Diffraction Powder Camera. E. Matuyama. (*J. Sci. Instruments*, 1955, **32**, June, 229–231). A high-temperature Debye-Scherrer camera is described, fitted with a small carbon-tube furnace for temperatures up to 1800° C. An example is given of the determination of the coefficient of thermal expansion of Ceylon graphite.—L. D. H.

The Theory of Eddy Current Testing in One (Not-So-Easy) Lesson. R. Hochschild. (*Non-Destructive Test.*, 1954, **12**, Sept.-Oct., 31–40). The physical basis of eddy current testing is briefly outlined, and the great versatility of the technique when properly applied is discussed. Some applications of the impedance analysis method to the measurement of conductivities, dimensions, thicknesses, and for flaw detection are discussed. The theory behind these test methods is presented. (1 reference).—P. M. C.

Tools of Inspection. R. H. Sparling. (*Non-Destructive Test.*, 1954, **12**, Sept.-Oct., 19–28). A detailed review is presented of non-destructive test methods which indicate the " quality " or soundness of metal parts.—P. M. C.

IRSID Committee for Studying High Temperature Metals, Including the High Temperature Properties of Steels Containing 0·6% Cr, 0·6% Mo; 2·25% Cr, 1% Mo; 22% Cr, 18% Ni Respectively. G. Delbart and A. Constant. (*Rev. Mét.*, 1954, **51**, Dec., 845–868). Since 1947, a committee set up by IRSID

has studied and defined the high temperature properties of French steels used in thermal power stations. This report gives the results obtained, in particular of tests on two low-alloy Cr–Mo steels and an austenitic Ni–Cr steel. The report also gives details of tests on the same steels by the steel manufacturers concerned.—G. E. D.

Note on the Use of Steels Subjected to Triaxial Stresses at High Temperature. W. Siegfried. (*Rev. Mét.*, 1955, **52**, Mar., 201–218). Extended tests at high temperatures were carried out on the creep resistance of various steels used in gears. Smooth and notched tubes made of these steels were used for the tests.—G. E. D.

Rynalloy Developed As Heat, Corrosion Resistant Cast Alloy for Aircraft. W. G. Hubbell. (*Western Metals*, 1953, **11**, July, 68–69). The development of a new cast alloy for use in ball and socket joints in aircraft exhaust systems is described. The alloy has a maximum service temperature of 1800° F and is sand cast at about 2350° F. The approximate composition is 20% nickel, not less than 1·8% chromium, silicon less than 6%, carbon less than 2½%, and sufficient manganese to counteract the sulphur—but not more than 1%.—P. M. C.

Graphitization in Steel. A. M. Hall. (*Mat. Methods*, 1954, **40**, Nov., 96–99). The failure-producing phenomenon of graphitization, which may occur in engineering steels operating in the 800–1200° F temperature range and in welded joints, is discussed. The avoidance of this trouble by chromium additions to the steel and by heat treatment of welds is described. (10 references).—P. M. C.

Selecting Metals for Supersonic Aircraft and Guided Missiles. S. L. Shaw. (*Mat. Methods*, 1954, **40**, Dec., 89–92). The problem of aerodynamic heating at supersonic speeds and its critical importance in the design of modern aircraft and rockets is discussed from the point of view of materials selection.

Creep Tests With Variation in Load at Constant Temperature.—A. Erra. (*Met. Ital.*, 1955, **47**, Feb., 53–62). [In Italian]. This paper describes an experimental study of creep when the metal is submitted to interrupted loading and overloading at constant temperature. The effects on creep of variable loading conditions are examined in detail, i.e. the effects of removing the load once, a number of times, loading cyclicly and for variable durations. Both experimental techniques and results are described in detail. (15 references).—M. D. J. B.

The Creep of Metals. E. Nygren. (*Jernkontorets Ann.*, 1955, **139**, (3), 182–204). [In Swedish]. The article reviews past attempts to obtain a theoretical explanation for the creep process. Research work on the creep in steel at room temperature is also discussed. (31 references).—G. G. K.

The Embrittlement of Steel by Hydrogen. Winifred A. Bell. (*Product Eng.*, 1955, **26**, Mar., 189–192). A brief review of the causes and effects of hydrogen embrittlement is given by the author. Pick-up can occur in steelmaking, welding, pickling, plating, and during corrosion. Loss of ductility and cracking at welds can be minimized by a critical cooling rate from about 300° C. The detection of embrittlement and conditions under which hydrogen can be removed are also discussed.—A. M. F.

High Alloy Steels containing Nitrogen. A. Semkowicz. (*Hutnik*, 1955, **22**, (1), 8–12). [In Polish].

Influence of Addition of Nickel on the Thermal Expansion, Rigidity Modulus and its Temperature Coefficient of the Alloys of Cobalt, Iron and Chromium, especially of Co-elinvar. I. Additions of 10 and 20% Nickel. H. Masumoto, H. Saitô and T. Kôno. (*Sci. Rep. Res. Inst. Tôhoku Univ.*, 1954, A, **6**, Dec., 529–538). A study of the addition of 10 and 20% nickel on the thermal expansion in range 10–50° C, on the rigidity modulus at 20° C and on its temperature coefficient in range 20–50° C, of co-elinvar is reported. The properties are similar to those of the ternary alloys, but the addition of nickel improved the mechanical properties.—J. G. W.

Study of the Action of Sulphur in Steels: Practical Applications. A. Paudrat (*Bull. Cercle Etudes Mét.*, 1954, **6**, Dec., 289–318). The modern theories of deformation and destruction of crystals by external forces are reviewed. Reference is made to single and polycrystalline materials and to the cutting of metals. The mechanism of the action of sulphur on modifying the properties of polycrystalline materials is then discussed. The conditions for the introduction of sulphur into alloy steels are reviewed together with industrial processes of sulphur nitriding and sulphur cementation. The results of practical tests on sulphur bearing steels are reported.

Impurities in Metals: Their Function in Providing some Important Properties. H. W. Greenwood. (*Metal Treatment and Drop Forging*, 1954, **21**, Dec., 571–573). The author discusses briefly the role of some impurities deliberately introduced into various metals to provide specific properties. Quoted examples include the use of lead to improve the machinability of steel, magnesium in the production of S.G. cast iron, and the improvement of magnetic and electrical properties, hardenability, and structure of many ferrous and non-ferrous metals by various other metallic impurities.

Effects of Titanium upon the Quality of High Chromium Steels. II. Effects of Titanium on Some Properties of 18% Chromium Steels. T. Murakami and T. Suzuki. (*Nippon Kinzoku Gakkai-Si*, 1951, **B**, **15**, Dec., 594–599). [In Japanese]. The properties of steels containing 0·03–0·27% C, 0–3·26% Ti and 18% Cr were studied. Self-hardening is not found at Ti >0·25% and increase of Ti content increases hardness. When Ti >2·5% temper-hardening is found. Approx. 0·25% Ti refines ferritic grain size and retards growth at high temp. Impact resistance increases markedly in the range 0·25–0·50% Ti.—K. E. J.

On Abridged Creep Tests on Materials Supplied for Boiler Construction. K. Wellinger and E. Keil. (*BWK*, 1955, **7**, Aug., 354–356). Tests on boiler plate at the operating temperature after annealing for some periods at higher temperatures, and tests at temperatures above that used in practice, showed them not to be dependable for assessing the performance of the material in prolonged use.—P. F.

Quality Steels in the United States. (*Usine Nouvelle*, 1955, July, 23–28). A table summarizes the essential differences between the two types of constructional steels, i.e. those that are tempered throughout, and surface tempered steels. Choice of type depends on the specific use, and examples are given. The characteristics of the two types are discussed and include mechanical resistance, elasticity, resilience, resistance to fatigue, and resistance to wear. Some testing methods are mentioned. It is indicated where French and American steels correspond.—T. E. D.

Austenitic Manganese Steel: Properties and Uses. T. H. Arnold. (*Iron Steel*, 1955, **28**, Mar., 95–97). The author outlines the characteristics of 1·2% C, 12½% Mn austenitic steel, and discusses the principles governing its heat treatment. Some important practical aspects relating to production and service conditions are explained.—G. F.

High-Strength, Low-Alloy Steels. C. M. Parker. (*Weld. Res. Council, Weld. Res. Council Bull. Series*, 1955, (20), Jan., 1–5). The mechanical properties, corrosion resistance and weldability of high strength, low alloy steels are discussed.—U.E.

Metallurgical Problems of Modern Steam Turbines and Alternators. F. Buckley. (*Eng. J.*, 1955, **38**, July, 919–928). The development of large forgings and castings of special steel to meet improvements in generating practice is outlined, with special reference to steel melting practice, stress-relieving heat-treatment, hardening and tempering, creep properties, alloy composition, jointed assemblies of different compositions, and casting practice. Inspection techniques are developing rapidly, and the uses of physical, ultrasonic, stability, magnetic and radiographic test methods are described.—K. E. J.

Production of Seamless Tubes in Special Steels. (*Aciers Fins Spéc. Franç.*, 1955, Mar., 43–44). A number of examples are given of special steels used in the manufacture of piping when certain particular qualities are required, e.g. high mechanical strength, resistance to corrosion, resistance to oxidation at high temperatures.—E. A. C.

New Viewpoints on Alloyed Constructional and Tool Steels. A. Michel. (*Metalen*, 1955, **10**, Apr. 30, 103–109). [In Dutch]. The author points to the growing exhaustion of present known sources of supply of alloying elements used in steelmaking. Co-operation is called for between steelmakers and consumers in order to prevent unnecessary consumption. The need for research is pointed out on new steels and their heat treatment. Some figures are given on the production and sources of supply of alloying elements and on the results of investigations in heat treatment of lower alloyed steels.

METALLOGRAPHY

The Solubility of Graphite in Manganese, Cobalt, and Nickel. E. T. Turkdogan, R. A. Hancock, and S. I. Herlitz. (*J. Iron Steel Inst.*, 1956, **182**, Mar., 274–277). [This issue].

Some Effects of Heat-treatment and Microstructure on the Transition Temperature of a 0·24% Carbon Steel. G. Burns and C. Judge. (*J. Iron Steel Inst.*, 1956, **182**, Mar., 292–300). [This issue].

Investigation of the Relationships between Physical Properties and Structure of Steel Wires. H. Schlacher. (*Berg. Hüttenmänn. Monatsh.*, 1955, **100**, May, 166–170). A variety of steel wires were subjected to a selection of heat treatments. An attempt is made to relate the resulting textures to strength properties.—E. C.

The Effect of Tempering on the Structure and Properties of Hardened 13% Chromium Steel. S. Heiskanen. (*Jernkontorets Ann.*, 1955, **139**, (6), 361–411). [In Swedish]. The paper discusses carbide reactions during tempering and describes the investigation of carbides by means of optical and electron microscopes and X-ray diffraction. The steel's low corrosion resistance at a certain stage in tempering could be explained by local cell formation, due principally to differences in chromium content, between the carbide/ferrite contact surface and the rest of the ferrite.—G. G. K.

The Amalgams as an Aid to Metallographic Research. F. Lihl. (*Z. Metallkunde*, 1955, **46**, June, 434–441). Attainment of equilibrium conditions in alloys, particularly if the components have widely different melting and boiling points, is more easily achieved in many cases by the use of amalgams. A highly-soluble amalgam such as zinc can combine easily with an iron amalgam, although iron is practically insoluble in mercury; if both components have low solubilities, as with chromium and iron, amalgams are obtained by electrolysing mixed solution of the metals on to mercury. By careful control of the conditions of electrolysis various alloys of Cr–Fe, Co–Fe, and Cd–Fe are obtained; the mercury is afterwards distilled off. Experimental conditions and results are discussed.—L. D. H.

On the Cu-Collodion Replica Method for Electron Microscopy. N. Takamoto and Y. Tanabe. (*Mem. Fac. Techn.*, 1954, (4), 235–240). A superior replica method to that using SiO_2, previously described, is reported; it is applicable to all forms of Fe and steel. Cu is deposited on to the surface electrolytically, and a negative replica stripped off. Collodion is then run on to this, and the Cu is dissolved out by acid from the composite article, to give a positive collodion replica. 60 min are required. The collodion replica is more durable than one of SiO_2; the method also has the advantage that the Cu replica can be examined immediately if desired.—K. E. J.

Studies on Graphite in Cast Iron by Electron Diffraction. I. Structure of Flake Graphite. K. Niwa and G. Shimaoka. (*Nippon Kinzoku Gakkai-Si*, 1955, **19**, Jan., 43–46). [In Japanese]. Single graphite flakes extracted from molten C- or Si-rich iron at 1200–1500° C have been studied; well-developed large flakes consist of single crystals, and thin, almost colourless flakes of two or three crystals. Details of the arrangement and structure of these crystals are given.

An Electron Diffraction Study to Determine the Oxide Form produced in Oxidised Molten Iron. N. Kayama. (*Waseda Univ., Rep. Castings Res. Lab.*, 1955, (6), 6–8). Blowing of molten Fe with O_2 produced values of SiO_2 (as inclusion) up to 0·05%, although the FeO content did not exceed 0·02%, as determined by the Cl_2 method. Evidence from electron-diffraction investigations showed that this SiO_2 existed as α-cristobalite, and did not arise from the breakdown of silicates.—K. E. J.

The Use of Formvar for Mounting Metallographic Specimens. J. V. Hardy and A. D. Hopkins. (*Metallurgia*, 1955, **51**, Apr., 209–211). The advantages of this method are described and details are given of the technique using normal metallographic apparatus.—B. G. B.

Examining Machined Surfaces by Interferometry—Comparison of Hobbed and Shaved Gear Teeth. J. Dyson. *Engineering*, 1955, **179**, Mar. 4, 274–276). A new interferometric method of examination offered to the tooth surfaces of an experimental pinion brings out clearly the difference in surface finish obtained by hobbing and shaving respectively.

Self-Oxidation for Obtaining Foil for the Electron Microscopy of Metals. I. Pfeiffer. (*Z. Metallkunde*, 1955, **46**, Aug., 569–573). A method is described for the production of oxide films for the examination of metal structures. The section is heated to 500° C in an oxidizing salt bath, or in air to a suitable temperature, and the film, 200–300 Å in thickness, is stripped by dipping in a solution of bromine in methanol.

Thermionic Emission Microscopy of Metals. II. Transformations in Plain Carbon Steels. R. D. Heidenreich. (*J. Appl. Phys.*, 1955, **26**, July, 879–889). A description is given of the direct observation of both A_3 and A_1 transformations at temperatures above about 625° C. Emission studies of the decomposition of austenite below the A_3 line indicate that the transformation is of a diffusionless type rather than carbon diffusion controlled. Emission images also indicate that the growth of a ferrite grain in austenite induces a recrystallization of the surrounding austenite. Studies below the A_1 line illustrate the formation of pearlite. Decompositions appear to take place in two steps—the formation of a metastable crystal which subsequently decomposes to the final ferrite–carbide products.—E. E. W.

A Microscope Hot Stage. H. A. Saller, R. F. Dickerson and R. J. Carlson. (*Metal Progress*, 1955, **67**, May, 105–108). A short account of a microscope hot stage suitable for heating specimens in a vacuum of 10^{-4} mm mercury and up to 1800° F is given. No special cooling device has been found necessary and normal objectives are used. A number of micrographs taken at high temperatures are reproduced.—B. G. B.

A High-Resolution Metallograph for Elevated Temperatures. J. E. Jenkins, D. R. Buchele and R. A. Long. (*Metal Progress*, 1955, **67**, May, 101–104). A description is given of a microscope hot stage and a special objective with a high resolving power and a long working distance. The specimen can be heated to 1800° F in vacuo or in an inert atmosphere. Examples of the application of the technique are given.—B. G. B.

Surface Studies of Solids by Total Reflection of X-Rays. L. G. Parratt. (*Phys. Rev.*, 1954, **95**, July 15, 359–369). A new method of studying certain structural properties of a mirror surface (10 to +100 angstroms deep) is described. Reflected X-ray intensity is plotted against glancing angle in the region of total reflection and an analysis is made of the shape of the curve.—E. E. W.

A Theory on Image Formation of Electron Microscope. R. Uyeda. (*J. Phys. Soc. Japan*, 1955, **10**, Apr., 256–264). The author's theory, which is based on the idea of building up an image of an object from images of the atoms constituting the object, suggests that the actual image consists of two ideal images, one showing phase contrast, the other contrast due to scattering absorption. The thoery explains bright and dark field images.—J. G. W.

A Universal Polishing Method. H. S. Cannon. (*Metal Progress*, 1955, **67**, Apr., 83–86). A standard procedure for polishing metallographic specimens is described in detail.

New Etching Agents for Certain Types of Ferritic, Semi-Ferritic, Martensitic and Austenitic Stainless Steels. G. Catella and C. Giometto. (*Met. Ital.*, 1955, **47**, Jan., 19–20). [In Italian]. The authors describe their search for new and more satisfactory etching agents particularly for stainless steels with low and very low carbon content and with chrome contents ranging from 12% to 18% (ferritic, semi-ferritic and austenitic) and for classic austenitic types of the 18/8 variety. The chemical compositions of two agents are given: the first containing ethyl alcohol, acetic, hydrochloric and picric acids; the second containing nitric, hydrochloric, fluoric and acetic acids.—M. D. J. B.

Apparatus for Mechanical Polishing of Micro-Samples Having Continuous Feed of Polishing Medium. K. Tikkanen, (*Jernkontorets Ann.*, 1955, **139**, (4), 265–268). [In Swedish]. This machine is similar in design to the apparatus evolved by L. Ekbom (*Jernkontorets Ann.*, 1954, **138**, (8), 490–494), and is equipped with a guide-ring containing up to six samples. The ring rotates slowly so that the samples pass repeatedly across the centre of the polishing table. The polishing medium is contained in a flask and can be gravity fed or forced out via a capillary tube by means of compressed air and regulated by varying the pressure.—G. G. K.

The Influence of Elements of Low Solubility on the Properties of Ferrite. W. P. Rees. (*Rev. Mét.*, 1955, **52**, May, 375–391). Very small amounts of carbon, nitrogen and oxygen have large effects on the properties of iron. If kept in solution by quenching, carbon gives strength to iron with loss of ductility. Similarly nitrogen gives strength but with marked reduction in grain boundary strength. Oxygen strongly weakens grain boundaries; this effect progressively increases as the oxygen content rises. The rate at which alloys are cooled from high temperatures controls the influence of these three elements.

On the Carbides in Manganese Steel. H. Yoshisaki. (*Sci. Rep. Res. Inst. Tôhoku Univ.*, 1954, A, **6**, Oct., 469–476).

Manganese steels in range 0–20% were investigated to elucidate whether the carbides are solid solutions of cementite and manganese carbide or chemical compounds thereof. In contrast to other investigators' conclusions no evidence of a chemical compound was obtained. The investigation was based on chemical analysis and thermal changes in the magnetic susceptibility of the samples.—J. G. W.

Rolling and Recrystallized Structures in Very Soft Steel Sheet. G. Pomey and C. Crussard. (*Rev. Mét.*, 1955, **52**, May, 401–417). The development of a new X-ray diffraction chamber has enabled IRSID to perfect a technique for accurately defining crystal orientations developed during hot-rolling, cold rolling and subsequent annealing. The study of various structures has led to a theoretical interpretation of the nature and formation of preferential orientations during rolling and recrystallization.—G. E. D.

Grains Within Grains. R. C. Gifkins. (*Australian Inst. Met.: Australasian Eng.*, 1955, Feb. 7, 41–50). The history of the discovery, and development of theories of origin, of various types of sub-grains is reviewed. The author deals with 'cells' formed during creep, sub-structures due to 'polygonization' and 'recrystallization in situ' in single crystals, macro-mosaic structures found in annealed metals, and the mosaic-type structures revealed by the electron microscope. (81 references).—P. M. C.

Dislocation Networks in Crystals. T. Suzuki and H. Suzuki. (*Sci. Rep. Res. Inst. Tôhoku Univ.*, 1954, A, **6**, Dec., 573–596). The conditions for existence and stability of dislocation networks were examined, giving an estimated dislocation density of 2×10^6 cm/cm^3 in annealed f.c.c. cubic crystal. Several observations concerning micro-creep could be accounted for in terms of hexagonal network in $\{113\}$ planes.—J. G. W.

An Isothermally Acting Calorimeter for Metallographic Investigations. J. Diehl and R. Braun. (*Z. Metallkunde*, 1955, **46**, June, 457–461). An improved differential isothermal calorimeter is described, which substantially shortens heating-up times without affecting the sensitivity of the method.—L. D. H.

On Metallographic Terminology: Oligocomponents and Oligoconstituents. A. Portevin. (*Usine Nouvelle*, 1955, July, 53). In the chemical composition of a metal as determined by analysis, the oligocomponents are the impurities and the small additions, e.g., aluminium, boron, etc., in steels. Oligoconstituents are the microscopic constitutents, such as inclusions in steel, as shown by physical analyses. Examples are given.

Strength of Metallic Materials as an Atomic Phenomenon. W. Kossel. (*Z.V. d. I*, 1955, **97**, June 11, 516–518). This is a popular exposition of the dislocation mechanism of deformation and crystal growth illustrated by analogue models.

Diffusion in Solid Metals and Alloys. G. P. Chattergee. (*Trans. Indian Inst. Met.*, 1953, **7**, 223–231). The author has attempted to improve Fick's Law by considering diffusion on the basis of potential and energy density. He has shown that the equation developed leads to all other diffusion equations. Some fundamental concepts and critical discussions are included.—D. H.

Electrolytic Migration of Carbon in Steel. P. Dayal and L. S. Darken. (*Trans. Indian Inst. Met.*, 1953, **7**, 241–250). A critical review is presented of the existing knowledge of the migration of carbon in austenite under the influence of direct current. The paper further deals with investigations carried out by the authors on S.A.E. 1045 and 1095 steels. The charge on the carbon ion has once more been proved to have a classical value of four (C^{4+}), and it has been shown that the phenomenon of electrolytic migration is superimposed by thermal diffusion, that is, migration of carbon from a cool to a hot zone.—D. H.

Electron-Diffraction and its Use in Metallurgy, particularly in the Metallography of Steel. F. Králík. (*Hutnické Listy*, 1955, **10**, (5), 262–266). [In Czech]. The principles and recent applications are reviewed. In a Ni–Cr–W steel prone to temper-brittleness application of the method yielded no positive results in relation to this behaviour.—P. F.

The Electron Microscope and Some of Its Industrial Applications. (*Machinery*, 1955, **86**, Apr. 15, 805–812).

Apparatus for Linear Microscopic Analysis. G. Molinder. (*Jernkontorets Ann.*, 1955, **139**, (3), 174–181). [In Swedish]. Constructed primarily for use with a Reichert MeA microscope this apparatus is equipped with three counters, permitting the determination of up to three constituents in any sample. Examples are given of its use and accuracy (maximum difference between chemical and linear analysis was 0·4% by volume). With trained operators the time taken for measuring a 1 mm length of sample varies between 5 and 20 minutes, depending on the appearance of the structure.—G. G. K.

A Microscope Hot Stage Using Electron Bombardment. G. J. Ogilvie and G. Brinson. (*J. Sci. Instruments*, 1955, **32**, May, 170–171). To obtain rapid and uniform heating of a metal specimen, the latter is supported in a vacuum between three pointed wires, two of which are thermocouple elements, and is heated by electron bombardment from a filament placed about $\frac{1}{8}$ in. from the specimen. A high rate of heating (rates much higher than 2000° C/min are said to be possible) and a low power input are features of the instrument.—L. D. H.

Electronic Metallography. L. Habraken. (*Rev. Soc. Roy. Belge Ing.*, 1955, Mar., 136–162). A comprehensive account of the use of the electron microscope for studying heat resisting ferritic alloy steels is presented. Details are given of the preparation of replicas of formvar. 134 optical or electron photomicrographs of various structures are reproduced. Details are given of the etching and replica techniques used on each. The application of X-ray diffraction methods in this field is also described.—B. G. B.

Application of Photocells in the Metallographic Investigation of Cast Iron. E. I. Egorov and Yu. S. Romashin. (*Liteinoe Proizvodstvo*, 1954, (9), 22–23). [In Russian]. To replace the estimation of the graphite-content of iron based on visual comparison with a standard micro-structure, a photoelectric arrangement has been developed. An outline of the proposed method is given.—S. K.

Structure Investigation of Steel by Means of a Low-Power Electron Microscope. E. Kinder. (*Arch. Eisenhüttenwesen*, 1955, **26**, Feb., 113–116). The low-power electron microscope developed by the author closes the gap between the optical microscope and the conventional electron microscope. The new microscope is equipped with electromagnetic lenses, operates at 50 kV and has a maximum resolution of 6 Å. Two magnifications are used, 650× and 3200×. The 650× magnification gives images of extremely high contrast—far superior to the optical microscope—these images can easily be enlarged up to 6000×. The low-power electron microscope is sufficient for most metallographic studies and is often, for special purposes, even superior to the conventional electron microscope. Excellent micrographs illustrate the paper.

On the Electron Microstructure of Slack Quenched Steel. T. Kawai, K. Ogawa and R. Kazui (*Sumitomo Metals*, 1954, **6**, Jan., 51–70). [In Japanese]. Studies were made on Jominy hardenability specimens of several carbon and low alloy steels, using the electron microscope and a replica technique of formvar-aluminium with chromium shadowing. The occurrence of proeutectoid ferrite, bainite and fine pearlite was demonstrated and formation mechanisms are proposed. (14 references).—K. E. J.

Recent Development in the Equipment for the Preparation and Examination of Metallographic Sections. R. Pusch. (*Stahl u. Eisen*, 1955, **75**, Mar. 24, 335–345). The progress made in the equipment for preparation and examination of metallographic sections in Germany during the last 20 years is described and illustrated.—T. G.

On Etching Steels with an Ionic Bombardment. I. N. Prilezhaeva, G. V. Spivak and M. I. Malkina. (*Zhurnal Tekhnicheskoi Fiziki*, 1954, **24**, (11), 2090–2096). [In Russian]. The use of the cathode atomisation as a universal etching agent was investigated. It was shown that the method is generally applicable and is superior to that of vacuum evaporation in that it can be carried out at any desired temperature of the specimen, and can be easily controlled. The variations in voltage lead to variations of relative atomisation of various components.—V. G.

Preparation of Steel Specimens for Micro-Examination by a Chemical Polishing Technique. L. W. Graham, J. P. Cranston and H. J. Axon. (*Research*, 1955, **8**, Apr., S20–S21). The chemical polishing of steel, previously suggested for macro-finishing, has been used to prepare surfaces for micro-examination at relatively high magnifications. The solution (freshly made for each specimen) was: oxalic acid (100 g/l.) 7 parts, H_2O_2 (100 vol.) 1 part, distilled water 20 parts; specimens prepared on grade "O" paper were immersed for 15 min at 35° C. By comparison with Nital etching, grain boundary carbide is strongly attacked, showing black, and, more important, previously-existing austenite boundaries are attacked. Polished specimens possess high corrosion resistance.

Some Aspects of the Oxidation of Iron at Low Oxygen Pressures, Allowing the Introduction of a Micrographic Method for Studying the γ-Iron Structure. J. Bardolle. (*Rev. Mét.*, 1954, **51**, Dec., 833–838). The first part of this paper extends the author's earlier work on the formation of iron protoxide nuclei, the number of nuclei per square unit being given as a function of the crystalline orientation of the metal. In the second part, the formation of iron protoxide nuclei provides a micrographic method for investigating the γ-iron structure. It is also considered possible to apply this method to the study of the $\alpha \rightleftharpoons \gamma$ transformation.—G. E. D.

Sigma Phase in Austenitic Stainless Steel Weldments. O. H. Henry, M. A. Cordori, and G. J. Fischer. (*Welding J.*, 1955, **34**, Feb., 75-s—81-s). The effect of sigma phase on tensile-impact properties of aged austenitic stainless steel welds in unnotched and notched conditions has been investigated. It was found that unnotched and notched AISI type 310 stainless steel welds were seriously embrittled at ageing temperatures of 1200°, 1350° and 1550° F, while AISI types 321 and 347 welds were embrittled at lower ageing temperatures of 1200° and 1350° F. The notch sensitivity of welds of the latter types increased at 1550° F. These results indicated that welds were more notch sensitive than base metal or heat-affected zones in all alloys. A solution anneal of 1950° F restored the toughness of material embrittled at 1550° F.

Metallographic Atlas of Iron and Steel Microstructures. (*J. Inf. Tech. Indust. Fonderie*, 1954, Nov.-Dec. Supplement; 1955, Jan., Supplement). Photographs showing the structure of different irons are reproduced. Details of the polishing and etching techniques used are given. The conventions used by the A.S.T.M. are employed and 8 of the photographs show standard variations in the size of graphite flakes from >100 mm to $<1{\cdot}5$ mm. The structures of grey, white, and malleable irons are shown, together with those of special irons which have been heat treated.—B. G. B.

A Three-Dimensional Face-Centred Cubic Model for the Study of Crystal Phenomena. P. R. Rowland. (*J. Inst. Metals*, 1954–55, **83**, June, 455–459). A model containing 10^3–10^4 lattice units can be rapidly and accurately assembled by pouring steel ball-bearings into a suitable transparent mould. The application of this model to the study of crystals is considered.—B. G. B.

Studies on Crystal Boundaries in Deformed Metals, Especially by Micro-hardness Measurements. H. E. Tuchschmid. (*Thesis No.* 2327, *Eidgenössische Technische Hochschule, Zürich*, 1954). After discussing the theory and problems to be dealt with, the author goes on to describe the apparatus and methods used. Cast aluminium alloy (Al-Mg 5) and hot rolled Fe–Cr18–Ni8 are used for the tests. The chromium-nickel steel shows a yield point related to the austenite transformation as started by deformation itself. Slip lines can be seen after low deformation intensities and pseudo-martensite after increased ones. This shows the different behaviour of central and boundary regions in crystals. Crystal boundary hardening is considered unlikely. The atomic arrangement of crystal boundaries is discussed and a study of the single crystal in the polycrystalline aggregate is recommended (124 references).—R. P.

The Influence of Various Elements on the Grain Size of Austenite of Medium Carbon Steel. I. S. Gaev and V. V. Poloznikov. (*Zhur. Tekhn. Fiz.*, 1955, **25**, (3), 529–533). [In Russian]. The relation getween the grain size of austenite of medium carbon steel with significant concentrations of: Mn, Cr, Ni, Al, V, Mo, Si, Cu and Co was investigated. It was shown that elements lowering Ac_1 point increase the grain size of austenite at a standard temperature of its appearance and vice versa. No similar relationship with Ac_3 point was found.—V. G.

Bonding Forces and Distortions in Martensite Crystals. V. K. Kritskaya, G. V. Kudryumov and N. M. Nodina. (*Zhur. Tekhn. Fiz.*, 1955, **25**, (2), 177–181). [In Russian]. In order to obtain more accurate data on bond strength in martensite lattice and its dependence on the amount of dissolved carbon, the relative intensity of X-ray lines at two temperatures (room and $-185°$ C) was measured using molybdenum radiation. It is concluded that the resistance of martensite to plastic deformation is due to lattice distortions caused by dissolved carbon atoms.—V. G.

Single Crystals without Dislocations. H. K. Hardy. (*Research*, 1955, **8**, Feb., 57–61). Some metals (e.g., Sn, Zn, Cd) can form ' whiskers ' by a natural growth process; these are effectively dislocation-free single crystals, and have strengths far exceeding those of the normally-prepared metals. ' Whiskers ' of pure Fe have been obtained, approx. 0·001 in. thick and 1 in. long; they may have tensile strengths up to 400 ton/in² and high corrosion resistance. If ' whiskers ' can be used as seeds for filaments, great possibilities exist for applications such as cables with high strength/wt. ratio for engineering, structural and electrical uses, etc.—K. E. J.

The Decomposition of Martensite in Iron-Carbon Alloys. G. Folke and E. Nygren. (*Jernkontorets Ann.*, 1955, **139**, (4), 250–264). [In Swedish]. X-ray diffraction methods were employed in this study of high purity iron–carbon alloys having carbon contents ranging from 0·15 to 1·42%, also a carbon steel with 0·68%, 0·21% Si and 0·78% Mn. Results agree with previous findings—that the decomposition of martensite can be divided into two stages, the first and third stages of tempering. During the first stage ε carbide is precipitated whilst during the third cementite is formed at the expense of the ε carbide.—G. G. K.

An Investigation of the Martensite Transformation in Steel. A. B. Alfimov and A. P. Gulyaev. (*Zhur. Tekhn. Fiz.*, 1955, **25**, (4), 680–686). [In Russian]. Using a thermomagnetic apparatus of high sensitivity, kinetics of austenite–martensite transformation during a continuous cooling and the influence of some factors (grain size, dimensions of specimen, etc.) on the temperature of the beginning of transformation (M) were investigated. Steel used for this investigation contained: 1·5% C, 0·76 Si, and 3·4% Mn. It was found that: cooling of a single austenite crystal even to the temperature of liquid nitrogen is not accompanied by martensite transformation, but under the influence of a mechanical action this transformation takes place in all cases—before and after cooling. Cooling of two or more austenite grains between which the ledeburite eutectic was not dissolved was always accompanied by the formation of martensite. It is therefore concluded that the transformation is caused by stresses formed due to differences in the thermal expansion coefficient of individual austenite regions being in contact with each other.—V. G.

The Influence of Cooling Velocity on the Kinetics of Austenite–Martensite Transformation. A. P. Gulyaev and A. P. Akshentseva. (*Zhur. Tekhn. Fiz.*, 1955, **25**, (2), 299–312). [In Russian]. The dependence of the kinetics of austenite–martensite transformation on cooling velocity in the martensite region as well as transformation during heating under isothermal conditions, stabilization and some other features of this transformation were studied. It is concluded that the experimental data obtained cannot be explained without acceptance of the major influence of stresses in the process of formation of martensite nuclei.—V. G.

On the Method of Determination of the Coefficient of Self-Diffusion of Iron in an Alloyed Austenite. M. A. Krishtal. (*Zhur. Tekhn. Fiz.*, 1955, **25**, (1), 144–148). [In Russian]. A new method of determining the coefficient of self-diffusion of iron in austenite based on the determination of the velocity of decarburization of graphitized cast iron is described. Using this method the dependence of the coefficient of self-diffusion of iron in Fe–C–Si and Fe–C–Ni alloys on the contents of Si and Ni and the temperature was established.—V. G.

Isothermal Transformation of Ferrite in High Alloy Austenitic–Ferritic Steels. E. M. Pivnik. (*Zhur. Tekhn. Fiz.*, 1955, **25**, (1), 135–143). [In Russian]. The results of experimental investigations of the ferrite transformation in a number of high alloy austenitic–ferritic steels (containing from 5 to 30% ferrite) during prolonged isothermal treatment at temperatures 650–850° C are described. Some special features of the kinetics of the formation of sigma phase are discussed.—V. G.

Contribution on the Theory of the Graphitization of White Iron and Steel. M. A. Krishtal. (*Liteinoe Proizvodstvo*, 1953, (6), 22–24). [In Russian]. The theories of rate-control of graphitization in white iron and steel are critically described. The theory that the rate-controlling factor is diffusion of carbon atoms is compared with that which considers movement of iron atoms to be the main factor. Equations are given from which the criterion which determines which factor will be rate-controlling is derived. It is suggested that for cast irons and high carbon steels, the graphitization rate is controlled by movement of iron atoms, and for low and medium carbon steels by diffusion of carbon.—S. K.

Orientation Between Austenite, Intermediate Phases and Martensite. W. Hofmann and G. Schuhmacher. (*Arch. Eisenhüttenwesen*, 1955, **26**, Feb., 99–104). Pole figures from a

steel with 1·6% Mn, 0·4% Cr, 0·99% V, 0·2% Si, and 0·86% C were established by a modified Kurdjumow–Sachs method. The orientation relationship between austenite and the ferrite of the intermediate phase was found to be

austenite (111) ‖ (011) ferrite

$[11\bar{2}] \parallel [01\bar{1}]$;

and that of austenite and martensite

austenite (111) ‖ (011) martensite

$[10\bar{1}] \parallel [1\bar{1}1]$

The results are in agreement with those obtained by other workers using metallographic methods.—T. G.

An X-Ray Investigation of Nitrided Layers of Carbon on Alloy Steels. M. Ya. Fuks and E. V. Aronson. (*Zhur. Tekhn. Fiz.*, 1954, **24**, (8), 1448–1454). [In Russian]. During nitriding of alloy steels the concentration of ϵ phase (Fe_2N) in the surface layer is always lower than in carbon steel. Silicon in steel influences the hardness and the depth of the distribution of nitride phases. The hardness of the nitrided layer of steels alloyed with chromium, molybdenum and vanadium is due to the hardness of the α phase, and in steels containing in addition silicon or aluminium is due to the hardness of nitride phases. X-ray data on the amount of ϵ phase in alloyed steels, its lattice parameters and the width of lines on the X-ray photographs support the view that the formation of finely dispersed nitrides of alloying elements considerably increases hardness.—V. G.

Graphitization of ϵ-Phase of Nitrided Layer of a High Carbon Steel. A. N. Serov. (*Zhur. Tekhn. Fiz.*, 1954, **24**, (10), 1788–1801). [In Russian]. It is shown that one of the main causes of a decrease in the surface hardness of a nitrided layer of a high carbon steel is graphitization in the ϵ phase. The graphitization process is accompanied by a considerable increase in volume which is similar to the growth of cast iron under similar temperature conditions. An intensification of the graphitization process in ϵ phase can be explained by the similarity of graphite and ε phase crystal structure. Alloying elements—chromium, manganese, molybdenum and tungsten—do not prevent but only retard the graphitization process.—V. G.

CORROSION

Corrosion Resistance of Some Austenitic Cr–Ni Steels of 18/8/Ti Composition. The Effect of Variation in Chemical Composition and Thermal Treatments. E. J. Heeley and A. T. Little. (*J. Iron Steel Inst.*, 1956, **182**, Mar., 241–255). [This issue].

Corrosion Protection for Aircraft Wheels. A. G. Thomson. (*Aircraft Eng.*, 1955, **27**, Mar., 78–79). An account is given of research work carried out in the U.S.A. concerning anti-corrosive lubricants for tapered roller bearings.—B. G. B.

The Prevention of Corrosion. (*British Iron and Steel Research Association Pamphlet*, 1954, 20 pp.). The problem of preventing the corrosion of iron and steel is considered in this booklet under the three headings: corrosion in air, corrosion in water, corrosion in soil. Corrosion may be alleviated by one or more of four methods and the relative usefulness of the various methods varies with the different environments. The methods considered include alterations in the corrosive medium, modifications in design, changes in the composition of the iron and steel and effective protective measures.—B. G. B.

Studies on Corrosion at Power Generation Stations Using Acid River Water. On the Corrosion of High Chromium Steels. H. Endo, S. Shimodaira and Y. Sawada. (*Nippon Kinzoku Gakkai-Si*, 1955, **19**, Jan., 4–8; 8–12). [In Japanese]. Additions of Cu, Mo and Ti to high-Cr steels are very effective in reducing corrosion by acid river water; 13% Cr steels containing appropriate Ti additions were scarcely corroded. The corrodibility of the steels was investigated in very dil. aq. acid solutions. Anodic polarization results are in agreement with observed corrosion-loss figures. Cathodic polarization results are also given.—K. E. J.

Electrochemical Properties of Alloys. V. On the Anodic Behaviour of Iron–Nickel Alloys in Sulphuric Acid Solution. S. Morioka and K. Sakiyama. (*Nippon Kinzoku Gakkai-Si*, 1955, **19**, Jan., 31–34). [In Japanese]. Increasing Ni content increases polarization and favours the onset of (temporary) passivation; the dissolution of alloys containing > 60% Ni does not cease entirely, however, on passivation. α-phase are less noble than γ-phase alloys. The formation of the ordered lattice Ni_3Fe reduces anodic dissolution.—K. E. J.

Study of the Intercrystalline Corrosion of an Austenitic Chromium–Nickel Steel. E. Brauns and G. Pier. (*Stahl u. Eisen*, 1955, **75**, May 5, 579–586). Current density/potential curves were taken of a stainless steel containing C 0·08, Cr 17·8 and Ni 7·6% which was rendered susceptible to intercrystalline corrosion by prolonged heating at 650° C. It was found that the grain boundaries became active at higher potentials than the grain face. Samples kept for long periods at this critical potential suffered intercrystalline corrosion in sulphuric acid. The authors studied also the effect of the redox potential in boiling $CuSO_4$–H_2SO_4 solutions in intercrystalline corrosion. The faster the redox potential arrives at equilibrium values which are within the critical potential range the stronger intercrystalline corrosion. Chemical analysis showed that this is caused by chromium impoverishment at the grain boundaries and their environment.—T. G.

Of What Use are Chemical Treatments in Controlling Corrosion. S. T. Powell and L. G. Von Lossberg. (*Mech. Eng.*, 1955, **77**, June, 495–498). In this brief review the authors evaluate the various chemical reagents used for corrective treatments of boiler water and feed water in steam generating plant. Attention is given to reagents used to remove oxygen or inhibit its effect on corrosion and miscellaneous inorganic and organic materials to minimize corrosive attack on metals. Internal softening and sludge conditioning are not discussed. D. H.

Corrosion by Heat-Transfer Liquids. M. R. Bigeon. (*Corrosion et Anti-corrosion*, 1955, **3**, May-June, 116–124). The materials considered do not necessarily remain in the liquid state while they are in use. They include saturated water vapour, superheated water, chlorinated hydrocarbons, polyglycols, mineral oils, organo-silicates, hydrocarbons and aromatic oxides, metals and molten salts. Maximum operating temperatures are given, and also data concerning their corrosive action.—T. E. D.

The Inhibiting Role of Long Chain Amines in the Petroleum Industry. B. Nauton. (*Corrosion et Anti-corrosion*, 1955, **3**, May-June, 113–115). Tests have been performed in a refinery in which amines with a high molecular weight were injected into the vapour line. The scheme is shown diagrammatically. The amount of corrosion was reduced. A monomolecular layer of the amine is probably absorbed on to the metal surface.—T. E. D.

The Formation of Protective Coatings and Steam Dissociation in Steel Tubes at High Temperatures. E. Ulrich. (*BWK*, 1955, **7**, June, 241–248). The amount of steam dissociated in steel tubes is a measure of the rate of their oxidation. The extent of the latter can be estimated from two measurements of hydrogen concentration in the steam, taken at various periods during the life of the tube. General formulae for the calculation of hydrogen evolution are derived, using Wagner's corrosion theory, and confirmed by the use of data given in the literature.—P. F.

Feedwater Treatment for Modern High Pressure Boilers. J. D. Yoder. (*Blast Furn. Steel Plant*, 1955, **43**, July, 772–776, 792). An account is presented of the application of ion exchange techniques to the treatment of water for use in high pressure boilers.—B. G. B.

Attention to Corrosion in the U.S.A. F. L. LaQue. (*Chem. Indust.*, 1955, Aug. 13, 1016–1025). The author describes the general activities of the main American technical societies and government agencies concerned with corrosion and its prevention. Some of the committees and sub-committees are enumerated in a useful appendix.—E. E. W.

Corrosion in Industry. N. S. Boas and S. A. Ambrose. (*J. Inst. Eng. Australia*, 1955, **27**, Apr.-May, 109–117). The modern theory of corrosion is briefly described and is followed by examples of the corrosion of external structures, marine equipment, power and chemical plant. General wastage, pitting, fretting, high temperature and stress corrosion, corrosion fatigue, and caustic embrittlement are briefly explained. Methods of preventing corrosion are discussed.—B. G. B.

Stress Corrosion in Stainless Steel. R. A. Lincoln. (*Yearbook Amer. Iron Steel Inst.*, 1954, 172–180). The author discusses failure of stainless steels by stress corrosion, or localized transcrystalline cracking occurring in only mildly corrosive conditions when tensile stresses are present in the surface of the steel. A number of examples are quoted, and it is noted that the non-hardenable types of stainless steels are relatively immune to this failure. Possible causes of the phenomenon are discussed.—G. F.

Stress Corrosion in High Tensile Wire. W. O. Everling. (*Yearbook Amer. Iron Steel Inst.*, 1954, 185–202). The author describes an extensive investigation into the failure by stress corrosion recently observed in the oil-tempered wire used in place of hard-drawn wire in pre-stressed concrete pipelines. The results indicate that high-strength steel wire is subject to stress corrosion failure only in the presence of nitrate solutions, oil-tempered wire being substantially more sensitive than hard-drawn wire. The phenomenon appears to occur only when general corrosion protection breaks down.

Some Aspects of Stress Corrosion Cracking. T. P. May. (*Yearbook Amer. Iron Steel Inst.*, 1954, 206–213). The author describes a number of examples of stress corrosion in steel and discusses possible mechanisms of the phenomenon. It is suggested that, in addition to electrochemical action, hydrogen plays an important part which justifies further research. Other aspects needing further study are the location of cathodes and the role of chloride ions.—G. F.

On the Mechanism of Rusting of Painted and Unprotected Iron. A. Bukowiecki. (*Schweiz. Arch. Wiss. Techn.*, 1955, **21**, Apr., 121–133). A survey is made of the mechanism of rusting of iron and its modification in the presence of paint layers. The action of inhibitors with and without paint protection is discussed and the paper, supported by an extensive bibliography, also deals with the protection afforded by paint coatings on galvanized or rusted surfaces.—J. R. P.

Filiform Corrosion—A New Form. A. Bukowiecki. (*Schweiz. Arch. Wiss. Techn.*, 1955, **21**, May, 165–168). The author discusses the mode of formation of filiform corrosion of iron surfaces underneath a layer of paint. The factors influencing the effect of this form of corrosion are also considered.

The Prevention of Metallic Corrosion by Inhibitors. A. Bukowiecki. (*Schweiz. Arch. Wiss. Techn.*, 1955, **21**, June, 169–186). The action of inhibitors in preventing metallic corrosion by neutral, alkali and acid reactions in the presence of water or by water contained in organic fluids is reviewed. (121 references).—J. R. P.

Protection of Metals and Productivity. D. R. Jaumandreu Marimon. (*Dyna*, 1955, **30**, June, 356–367). [In Spanish]. A plea is made for a greater appreciation of the enormous cost to the country of corrosion and abrasion. The factors which contribute to a total cost—estimated to be of the order of 8000 million pesetas per annum—are reviewed. Economic methods of protection are reviewed, and the advantageous effects of these upon productivity are discussed.—P. S.

On the Kinetics of the Anodic Passivation of Iron. S. Feliu and M. Serra. (*Anal. Fis. Quím.*, 1955, **51B**, June, 395–400; 401–408). [In Spanish]. These two papers report a study of the anodic passivation of iron in solutions which contain (a) oxalic acid or sodium oxalate, (b) acetic acid or sodium acetate, (c) sulphate and chloride ions, and thus extend the authors' previous work. The relationship of the current density i to the time t necessary for passivation is such that $i\sqrt{t}$ is constant over a wide range of current densities except under certain conditions, i.e. it is a constant in solutions containing oxalate ions only; t is constant over a range of low current densities in type (c) solutions but the law changes to $i\sqrt{t}$ at higher current densities. There is also a critical current density for solutions of type (b) necessary for passivity, and this is greatly influenced by the composition of the solution and its pH value. The passivation kinetics are explained by a consideration of the electro-chemical behaviour of the ions in solution, and the tendency towards insoluble film formations at the anode surface. (10 references).—P. S.

Inter- and Transcrystalline Corrosion audits Testing. F. C. Althof. (*Metall*, 1955, **9**, Feb., 110–120). Following a review of knowledge concerning inter- and transcrystalline corrosion, methods of corrosion testing are discussed, classified and tabulated according to material, function of test, corrosive testing medium and procedure, and reference to literature. The range of metals and alloys surveyed includes straight carbon and high alloy (non-rusting) steels.—J. G. W.

High Temperature Corrosion Data. (*Corrosion*, 1955, **11**, May, 59–63). N.A.C.E. Technical Unit Committee T–5B. Corrosion rates are listed as 'high', 'moderate' or 'low' for 31 materials in over sixty corrosive media at temperatures above 400° F. The materials include mild steel; cast iron; Ni-resist, 4–6 Cr–Mo and 21 Cr–4Al steels; three Cr steels (12, 17 and 27 Cr); three Cr–Ni steels (18–8, 25–12, and 25–20); two Cr–Ni–Mo steels (18–12–2 and 18–12–3); silicon iron and Durimet 20.—J. F. S.

Economic Considerations in Pipe Line Corrosion Control. L. G. Sharpe. (*Corrosion*, 1955, **11**, May, 45–58). Three major problems encountered in pipeline operation are discussed from the point of view of relative costs of various corrosion prevention methods. These are (i) internal corrosion in sour crude oil pipe lines, (ii) external corrosion of oil storage tank bottoms, and (iii) external protection of pipes by coatings and/or cathodic protection. For (i), concrete linings and inhibitors are preferred on larger pipes while concrete linings, plastic linings and plastic pipes are preferred for smaller diameters. All methods of protection provide major savings. For (ii), it is suggested that ample cathodic protection applied to all tank bottoms is cheaper than selective protection. For (iii), coatings are an essential adjunct to cathodic protection but need not be perfect for best economy. There follows a discussion and reply.—J. F. S.

Laboratory Development of Corrosion Inhibitors. L. W. Jones and Jack P. Barrett. (*Corrosion*, 1955, **11**, May, 35–38). A new series of inhibitors has been developed which is effective in the oil producing industries even in the presence of hydrogen sulphide and oxygen together. Its development is traced through the laboratory tests and field trials.—J. F. S.

Corrosion of Metals by Acid River Water. I. Theoretical Consideration of Corrosion of Iron and Steel. S. Shimodaira. (*Sci. Rep. Res. Inst. Tôhoku Univ.*, 1954, A, **6**, Oct., 431–445). A theoretical study of the corrosion of iron and steels by acid river water showed it to be either cathodic or mixed. In stagnant water, the diffusion of dissolved oxygen determined the reaction rate above pH 3·6 and the discharge of hydrogen ions below pH 3·6. In flowing water, the roles of these two factors were reversed with pH 4·1 as the critical level. The corrosion rate was proportional to the seven-eighths power of the velocity of flow. Stainless steels might be passivated in high-speed flow. The 'moto-electrode' effect was shown to be important.—J. G. W.

Russian Work on Resistance to Heat and Corrosion of Surface-Alloyed Steels. W. G. Cass. (*Corrosion Techn.*, 1955, **2**, June, 197–198). This is a review in English of an article by G. N. Dubinin (*Vestnik Mashin*, 1953, **33(8)**, 74–79). Special attention has been given in Russia to aluminium that imparts stability to plain carbon steels up to about 950° C. Other elements used are Si, Cr, or combinations of Cr and Al, Al and Si or Cr, etc. (23 references).—L. E. W.

Corrosion in the Tannery. F. E. Humphreys. (*Corrosion Techn.*, 1955, **2**, June, 174–176). The author describes the metals used for different tanning processes and also methods of protecting structural steelwork of the tannery. (21 references).

Magnesium Anodes in a Chemical Plant. O. Osborne. (*Corrosion Techn.*, 1955, **2**, June, 170–173). This article describes some of the cathodic protection studies made by the author.—L. E. W.

How Carefully Do You Control Your Salt Spray Tests. H. A. Holden. (*Corrosion Techn.*, 1955, **2**, May, 157–159, 163). The author surveys briefly the 'salt-fog' testing method in which specimens or actual articles are contained in a closed cabinet and exposed to the 'fog' produced by the atomization of a salt solution. A comparision is made of the techniques used and the ASTM tentative specification B117–49T is discussed. (14 references).—L. E. W.

Preparation of Metals for Painting. R. E. Shaw. (*Corrosion Techn.*, 1955, **2**, May, 136–142). This comprehensive article surveys present-day techniques in the preparation of metals for painting and includes sections on the testing of results, economics and design considerations.—L. E. W.

Salt Spray Passes Acid Test. (*Steel*, 1955, **136**, May 9, 92–94). The ASTM 5% salt spray test, with the solution brought to pH 3·2 with acetic acid is recommended as a guide to atmospheric corrosion of plated metal.—D. L. C. P.

Collective Study of Under Water Paints. H. W. Talen. (*Peintures, Pigments, Vernis*, 1955, **31**, May, 425–432). A short account is presented of work carried out in Holland on this subject. Various compositions of anticorrosion and antifouling paints which have been proposed are reviewed.

The Corrosion and Marine Fouling Laboratories of C.R.E.O. M. W. Romanovsky. (*Peintures, Pigments, Vernis*, 1955, **31**, May, 433–435). A brief account of the activities of this French oceanographic research organization is given.—B.G.B.

Corrosion and Protection of Painted Steel Immersed in Sea Water. G. Dechaux and E. Segol. (*Peintures, Pigments, Vernis*, 1955, **31**, May, 436–446). The mode of corrosion of a steel sheet covered imperfectly by paint and immersed in sea

water s discussed. The formation of anodic and cathodic regions during corrosion is studied. The effect of alteration of the *p*H of sea water on the rate of corrosion of specimens is reported and the conditions for passivation determined. The electrical resistance and capacity of paint films on specimens immersed in sea water have been measured.—B. G. B.

Some Practical Aspects of Coatings for Submerged Parts of Merchant Ships. H. C. Fletcher and A. Partington. (*Peintures, Pigments, Vernis*, 1955, **31**, May, 411–424). An account is given of the activities of the Research Organization of Ship's Composition Manufacturers (ROSCM). The causes of underwater corrosion are first considered and suitable methods for protection are examined. (19 references).—B. G. B.

Corrosion Resistance Properties of Electrolytic Coatings of Tin Alloys. P. Déhais. (*Métaux–Corrosion–Indust.*, 1955, **30**, May, 185–191). The results of corrosion tests in salt solutions of coatings of tin–zinc alloy on steel are given. A coating of 75% tin, 25% zinc was found to give the best protection and this composition is now in commercial use. Electrolytic coatings of tin–nickel alloy containing 65% tin have also been studied. These coatings take a high polish and have good resistance to sulphurous atmospheres. For satisfactory resistance to corrosion the tin–nickel coatings should be deposited on a coating of copper.—B. G. B.

Acid-Resisting Steels For the Production of Acetic Acid. L. Wetternik and H. Zitter. (*Werkstoffe Korrosion*, 1955, **5**, June, 282–287). When exposed to boiling acetic acid, a 27/4·5/1·5 Cr–Ni–Mo steel, which is harder but only partly austenitic, proved much more resistant than 18/8/2 Cr–Ni–Mo steel, as generally used for plant producing vinegar.—J. C. H.

Surface Protection of Engineering Materials in Course of Fabrication. B. Van der Bruggen. (*Werkstoffe Korrosion*, 1955, **5**, May, 223–227). The labour involved in applying a protective coating to cast iron used for the manufacture of machinery is reduced by about 80% if the coating is put on to the parts immediately they have been received from the foundry, instead of waiting until they have been machined.—J. C. H.

Effect of Chemical Surface Treatment on the Resistance of Carbon Steels and Special Steels to Corrosion and Heating. W. Köhler. (*Werkstoffe Korrosion*, 1955, **5**, May, 228–236). (75 references).

Phenomena Resembling Stress Corrosion Caused by Diffusing Hydrogen in Steel under Stress. I. Class. (*Werkstoffe Korrosion*, 1955, **5**, May, 237–245). As a contribution to knowledge of transcrystalline stress corrosion in ferritic and austenitic steels, the author describes cases of cracking, mostly transcrystalline, in tubes of ordinary and alloy steels used in high pressure plant. He considers the cracking to be initiated by slight surface corrosion of the steel by substances such as hydrogen sulphide. This corrosion results in the formation of free hydrogen, which diffuses into the steel and ultimately produces very high local internal pressures.—J. C. H.

Metallurgical Aspects of Dry Corrosion. L. B. Pfeil. (*Chem. Indust.*, 1955, Feb. 26, 208–218). The author reviews the progress of the subject over 25 years, with emphasis on the practical aspects. Heat-resisting metals and alloys—especially those based on iron, cobalt and nickel—are discussed. The problems of attack by sulphur-bearing atmospheres and by fuel ash are indicated, together with 'green-rot' attack. Scale adhesion is also discussed. (36 references).—E. E. W.

Economic Cathodic Protection. B. G. Brand. (*Battelle Tech. Rev.*, 1955, **4**, May, 49–52). A short account of this form of protection is given. It can be made even more effective and economical by the use of organic coatings to limit the cathodically protected area.—B. G. B.

Some Corrosion Inhibitors—A Reference List. (*Corrosion*, 1955, **11**, Apr., 65–67). N.A.C.E. Technical Unit Committee T–3A on Corrosion Inhibitors. Names, sources of supply, where useful and with what metals, together with literature references are given for 68 inhibitors and groups of inhibitors.

Corrosion of Steel in Boilers—Attack by Dissolved Oxygen. H. A. Grabowski. (*Trans. Amer. Soc. Mech. Eng.*, 1955, **77**, May, 433–448). The various theories proposed to explain the severe, although local corrosion of furnace-wall tubes in high-pressure boilers are reviewed. Results of field tests of the oxygen record in utility-type boilers are presented. Metallurgical significance of failures and welding techniques are fully discussed.—D. H.

Action of Boiler Water on Steel—Attack by Bonded Oxygen. C. E. Kaufman, W. H. Trautman and W. R. Schnarrenberger. (*Trans. Amer. Soc. Mech. Eng.*, 1955, **77**, May, 423–432). The authors discuss the attack by boiler water on steel of boilers and superheaters. Emphasis is placed on the oxidizing power of water even in the absence of dissolved oxygen. Examples of damage caused by bonded oxygen have been explored by metallographic study of typical damaged specimens. Laboratory results show that hydrogen embrittlement can be produced and controlled in test bombs. It is concluded that the power of bonded oxygen of water to react with steel can be curbed by making both chemical and physical changes in boiler systems.—D. H.

Influence of Fine Particles on Corrosion of Economizer and Air-Preheater Surfaces by Flue Gases.—P. Hodson. (*Trans. Amer. Soc. Mech. Eng.*, 1955, **77**, Apr., 279–286). Flue gas corrosion is caused by the interaction of sulphur compounds and water vapour to form sulphuric acid which then attacks the iron of the economizer and air-preheater surfaces. The author discusses the effect of fine particles (less than 10 μ) of ash and unburned fuel in increasing the rates of these reactions and thus increasing corrosion.—D. H.

The Use of Additives for the Prevention of Low-Temperature Corrosion in Oil Fired Steam-Generating Units. E. C. Huge and E. C. Piotter. (*Trans. Amer. Soc. Mech. Eng.*, 1955, **77**, Apr., 267–278). Various methods of combatting low temperature corrosion of oil-fired steam generating units are discussed with particular reference to the use of dolomite as an additive to the fuel oil. This corrosion is caused by the presence of sulphur trioxide in the combustion gases unifying with the water vapour and forming sulphuric acid on the cooler surfaces of the unit. Results of actual operating experience are presented which show that additives are effective in reducing the amount of acid formed and in some cases practically eliminating cold-end pluggage and corrosion. The economics of their use and the design of units for higher efficiency when using additives are also discussed.—D. H.

Case History of Failure of Marine Boiler Tubes by Stress Corrosion Cracking. R. D. Barer. (*Corrosion*, 1955, **11**, Apr., 18–24). Details are given of an example of stress corrosion cracking of steel boiler tubes in the absence of seams or crevices to concentrate the boiler solution. It is shown that the attainment of the necessary caustic concentration and stress levels is feasible although it rarely occurs. The mechanism of caustic cracking is reviewed and practical measures for its avoidance are suggested.—J. F. S.

Corrosion by Valve Packing. L. M. Rasmussen. (*Corrosion*, 1955, **11**, Apr., 25–30). Corrosion encountered in stainless steel valves after hydrostatic testing is due to soluble corrosive agents in the asbestos braid and graphite lubricant. Tests showed that corrosion is reduced by treating the packing, selection of grade of graphite, addition of inhibitors, or by introduction of sacrificial metals as a washer or as powders. J. F. S.

Current Requirements for Cathodic Protection of Pipe Lines. M. E. Parker. (*Corrosion*, 1955, **11**, Apr., 52–58). The author discusses how to protect, most effectively and economically, a buried pipe line. Methods suggested of determining the best scheme are (i) placing the line under full protection using temporary trial installations; (ii) placing the line under partial temporary protection, and calculating the requirements for complete protection; (iii) measuring the electrical characteristics of the line and calculating how it would behave under various proposed systems and, (iv) estimating empirically what is required from coating and soil data, etc. The last method is important since it can precede and shorten the others. Field tests must not begin until the line has been electrically integrated and isolated from other especially large installations. The recommended method involves (i) determination of static potentials, (ii) application of current at one selected point, then others and measurement of potentials with current 'on' and 'off', (iii) graphical determination of the polarization/driving-voltage curves and the attenuation of driving voltage, (iv) calculation of effects of combining current from several points until best system is found. The techniques of potential measurement, positioning of half-cell, ways of plotting data, and criteria of protection are also discussed.

Cell Currents and Potentials. Robert Pope. (*Corrosion*, 1955, **11**, Apr., 59–64). Cell action is explained to electrical engineers by analogous equivalent circuits and polarization diagrams. The criteria of complete protection are defined with reference to a theoretical corrosion cell. The criteria of minimum cathodic protection are discussed and related to various electrical conditions.—J. F. S.

Research on Tin Corrosion and on Its Protective Action on Iron. G. Bianchi. (*Met. Ital.*, 1955, **47**, May, 216–219). [In Italian]. The physico-chemical aspects of the corrosion of iron when protected by tin are discussed. Research techniques and experimental results are described.—M. D. J. B.

Observations on the Corrosion Properties of Tinplate. W. E. Hoare. (*Met. Ital.*, 1955, **47**, May, 220–221). [In Italian]. The properties of tinplate as a material for packaging are discussed. Experimental techniques for studying the behaviour of tinplate when submitted to corrosive environment are described and the results are discussed. (7 references).

Polarographic and Colorimetric Determination of Tin and Iron in The Study of Tinplate Corrosion. F. Felloni. (*Met. Ital.*, 1955, **47**, May, 222–223). [In Italian]. It is suggested that the quantities of iron and tin which have passed into solution by various means of corrosion should provide a good measure of the electro-chemical behaviour of the two metals. The author suggests that a good method of establishing the corrosive behaviour of tinplate is to determine the quantities of one of the metals in relation to the other by using the polarographic method for determining the tin and the colorometric method for the iron. (11 references).—M. D. J. B.

ANALYSIS

The Absorptiometric Determination of Copper in Iron and Steel. B.I.S.R.A. Methods of Analysis Committee. (*J. Iron Steel Inst.*, 1956, **182**, Mar., 301–303). [This issue].

The Isolation of Silicon Nitride from a Commercial 4% Silicon Steel. H. A. Sloman. (*J. Iron Steel Inst.*, 1956, **182**, Mar., 307). [This issue].

Influences of Arsenic on the Determination of Various Elements in Iron and Steel. III. Studies on the Determination of Titanium and Molybdenum. H. Goto and S. Watanabe. (*Nippon Kinzoku Gakkai-Si*, 1955, **19**, Jan., 12–14). [In Japanese]. In the determination of Ti, As produces high results in the volumetric method using Zn amalgam and ferric sulphate solution when Ti is not separated, but does not interfere when Ti is separated as phosphate or in the gravimetric method. For Mo, As produces high results in the volumetric method based on separation by NaOH, but does not affect the gravimetric method—K. E. J.

Rapid Determination of Small Amounts of Carbon in Iron and Steel. II. T. Ikegami, O. Kamimori and A. Amano. (*Tetsu to Hagane*, 1955, **41**, June, 621–626). [In Japanese]. The best conditions for determining $<0{\cdot}05\%$ of C by a Co_2-absorption method include the use of a helical absorption tube of total length approx. 1·6 m and N/100 NaOH or KOH solutions. $0{\cdot}01$–$0{\cdot}05\%$ of C can be determined within $\pm$ $0{\cdot}001\%$ in 17–23 min, and the method is recommended for plant control purposes.—K. E. J.

On Segregation of Castings by Spectrographic Analysis. II. K. Yasuda and K. Amano. (*Waseda Univ., Rep. Castings Res. Lab.*, 1955, (6), 49–54). The segregation of Mg, Si and Mn was investigated in wedge-shaped castings of ordinary graphite iron; on these the " V "- and " W "-shaped segregation patterns, associated respectively with small- or large-diameter ingots, could be reproduced. Segregation became irregular in the region of shrinkage cavities, Si becoming higher and Mg and Mn lower. Si and Mn contents increased when slag inclusions were encountered. In the lower regions of the casting the local segregation is small, but the effects of cooling can be followed.—K. E. J.

On the Reliability of Ladle Analysis. O. Fujii. (*Tetsu to Hagane*, 1955, **41**, June, 626–631). [In Japanese]. A statistical analysis is presented of the degree of confidence which can be placed in the ladle sample as representing the chemical composition of the O.H. charge.—K. E. J.

On the Rapid Washing of Silica with Ammonium Tartrate Solution in the Determination of Si in Steel and Iron. S. Kiriyama, T. Masuo and H. Hoketsu. (*Sumitomo Metals*, 1955, **7**, Apr., 104–108). [In Japanese]. Time spent in washing SiO_2 can be reduced if ammonium tartrate solution is used instead of HCl. Details of the most suitable concentrations and number of washings are given.—K. E. J.

On a Rapid Method for Determination of Co in High Cobalt Alloy Steel. E. Dotani. (*Sumitomo Metals*, 1955, **7**, Apr., 97–103). [In Japanese]. A procedure is described for the rapid determination of Co within 1%, based on volumetric treatment with ferrous ammonium sulphate and titration with $KMnO_4$.—K. E. J.

Application of Spectrochemical Methods for Rapid Routine Analysis in an Integrated Iron and Steel Works. H. Hartleif and H. Kornfeld. (*Stahl u. Eisen*, 1955, **75**, May 5, 587–590). The authors describe the experience obtained with a Carl-Zeiss, Jena, Spectrograph Qu24 in the rapid routine analysis in a steelworks. In round-the-clock operation of the spectrograph about 180 samples can be analysed. The samples are specially prepared and then sparked and the films are evaluated every hour. Only two operators are needed. Agreement between spectrochemical and conventional chemical analysis is excellent, the spectrographic values for nickel are given $0{\cdot}01$–$0{\cdot}02\%$ too high, whereas those for aluminium are too low by about $0{\cdot}01\%$. The method is used in the present case for small contents only of the following metals: Al, Cr, Cu, Mn, Ni, Si, and V.—T. G.

The Simultaneous Determination of Vanadium and Molybdenum by Pyrocatechol. V. Patrovský. (*Chem. Listy*, 1955, **49**, (6), 854–857). [In Czech].

A Simple Instrument for Determining Hydrogen in Steel. M. Šícha. (*Hutnické Listy*, 1955, **10**, (8), 479–483). [In Czech]. A cheap, portable, apparatus suitable for works determinations of hydrogen is described, and the manner of using it discussed. About 80–100 should be necessary to complete a determination on one sample.—P. F.

Determination of Tin in Ferro-Molybdenum. O. Běloh-lávek and J. Vobora. (*Hutnické Listy*, 1955, **10**, (6), 354–355). [In Czech].

Carbon in Steel: A Review of Physical Methods of Determination. A. P. H. Jennings. (*Iron Steel*, 1955, **28**, May 11, 208–210). The author reviews a number of physical methods developed for the rapid determination of the carbon content of steel. Magnetic methods using the properties of saturation, coercivity, and permeability are first described and compared, and attention is also given to electrical, electromagnetic, and mechanical methods. It is considered that none of the methods will give better results than the magnetic methods which have proved satisfactory in practice, with an accuracy of $\pm$ $0{\cdot}02\%$ and a determination time of 2–3 min.—G. F.

Determination of Oxygen in Steel by the Vacuum Fusion Method. S. Yonezaki. (*Sci. Rep. Res. Inst. Tôhoku Univ.*, 1953, **37**, Apr., 89–95). The method, which is a modification of the Gakushin Hydrogen Reduction method, is described together with the vacuum fusion apparatus developed for routine works analysis. Results of works investigations are quoted, from which the author concludes that the oxygen content varies too much from ingot to ingot for a mean bath value to be deduced; that top-poured rimming steel ingots contain more oxygen than bottom-poured ones; that the oxygen content of the metal in a basic O.H. furnace is always greater than the value calculated from the equilibrium; and that the oxygen content increases in parallel with the carbon-oxygen equilibrium line during refining.—J. G. W.

Determination of Manganese in Iron, Steel and Ferromanganese (II). Determination of Manganese in Ferromanganese. H. Gotô and S. Watanabe. (*Sci. Rep. Res. Inst. Tôhoku Univ.*, 1955, A, **7**, Feb., 17–22). A study of the ferrous-permanganate method, after oxidation with persulphate, of determining manganese in ferromanganese is reported. A suitable analytical procedure, tested on several samples, and based on findings, is recommended.—J. G. W.

Photometric Determination of Aluminium in Steel. K. Wacykiewicz. (*Prace Inst. Ministerstwa Hutnictwa*, 1955, **7**, (1), 35–42). [In Polish]. A method for the determination of aluminium content up to $0{\cdot}05\%$ in steels using eriochromecyanine is described.—V. G.

A Spectrographic Method for the Quantitative Determination of Magnesium in Nodular Iron. Z. Czajkowna. (*Przeglad Odlewnictwa*, 1955, **5**, (2), 46–50). [In Polish]. A spectrographic method, suitable for the determination of the magnesium content of iron within the limits of $0{\cdot}03\%$ to $0{\cdot}5\%$ with an accuracy of $\pm$ 13% is described.—V. G.

The Photometric Determination of Cobalt by Extraction with β-Nitroso-α-Naphthol. A. Claassen and A. Daamen. (*Anal. Chim. Acta*, 1955, **12**, June, 547–553). [In English]. A method is described, applicable to the determination of cobalt in steel, in which Co is extracted with chloroform as the β-nitroso-α-naphthol complex. An acid solution of the sample is treated with sufficient sodium citrate to keep di- and trivalent metals in solution, and the *p*H adjusted to 3–4 with HCl or NaOH. H_2O_2 is added to prevent reduction of Fe^{3+}, and a 1% acetic acid solution of the reagent added, the

complex extracted with chloroform, and the Co determined photometrically at 530 mμ.—L. D. H.

The Rapid Estimation of Titanium in Alloy Steel. W. F. Pickering. (*Anal. Chim. Acta*, 1955, **12**, June, 572–576). [In English]. A method is described for the determination of Ti in alloy steels. The sample is dissolved in perchloric acid, a solution of E.D.T.A. in 5*N* NaOH added, and the solution made ammoniacal. Ti is precipitated, the process being accelerated by the addition of magnesium sulphate and the precipitate dissolved in sulphuric acid and determined colorimetrically after the addition of hydrogen peroxide. Time of analysis ½ hour, relative error ± 5%.—L. D. H.

Analytical Chemistry Fosters Progress in Steelmaking. C. M. Parker. (*Weld. Res. Council, Weld. Res. Council Bull. Series*, 1955 (20), Jan., 6–12). The role played by analytical chemistry in the development of new raw material sources and products is discussed.—U. E.

Practical Results of the Metallurgical Applications of Spectrographic Analysis. V. Mathieu, A. Hans, B. Rosen and A. Hannick. (*Rev. Mét.*, 1954, **51**, Dec., 824–832). Examples of routine applications of direct spectrographic analysis to the control of metallurgical production—G. E. D.

The Separation of Small Amounts of Phosphoric Acid from Iron and Vanadium by Distribution and Ion Exchange. W. Fischer, R. Paul and H.-J. Abendroth. (*Anal. Chim. Acta*, 1955, **13**, July, 38–45). [In German]. The effect of the *p*H value and the concentrations of P_2O_5 and MoO_4 on the extraction of P_2O_5 as the molybdate complex is investigated, ethyl acetate being used to extract the complex from aqueous acid solution. Separation can be achieved in the presence of a large excess of Fe and V under certain conditions. A cation exchanger was also used successfully to separate small amounts of P_2O_5 from Fe^{2+} and V^{4+}.—L. D. H.

Control and Perfection of a New Method For The Estimation of Tantalum and Niobium. B. E. Jaboulay. (*Chim. Analy.*, 1955, **37**, June, 198–200). A method for particular application to the analysis of iron alloys for tantalum and niobium is described. It is based on the fact that when a suitably prepared solution of tantalum in HCl/H_2O_2 is heated to near 100° C, a flocculent white ppt. is formed, whereas a similarly prepared and treated solution of niobium does not give a ppt., as perniobic acid is stable at 100° C in the presence of H_2O_2.—T. E. D.

Contribution to the Determination of Boron in Steels. L. Lenard and C. Dussart. (*Chim. Analy.*, 1955, **37**, June, 207–211). The method described is based on the colorimetric reaction between quinalizarine and boric acid. Two previously described methods, one for ordinary carbon–boron steels, evolved by Rudolph and Flickinger, and the other for alloy steels containing nickel–chromium–molybdenum–boron, developed by Weinberg, Proctor and Milner, are examined, and a number of difficulties are resolved.—T. E. D.

The Determination of Aluminium in Alloy Steels (Cryolite Method. R. Leo and G. König. (*Giessereitechnik*, 1955, **1**, Mar., 26–28). The method developed by J. W. Tananajew and P. H. Jakowlew has been investigated experimentally. Aluminium, chromium, manganese, nickel, copper, molybdenum, tungsten, and titanium can be accurately determined with commendable speed, by this method.—L. J. L.

The Development of a Rapid Method for the Determination of Chromium in Steel. K. Merz and M. Schnabel. (*Metallurgie*, 1955, **5**, May, 163–165). A more rapid method for the determination of chromium than the Philips method has been developed experimentally to enable monitoring of chromium content during the melting process.—L. J. L.

Determination of the Sulphur Content of Cast Iron by Combustion with Oxygen. J. Scheunemann, Jun., (*Giessereitechnik*, 1955, **1**, Mar., 29–30). The combustion method of sulphur determination has been found rapid, relatively simple, and entirely satisfactory.—L. J. L.

A Solution Method of Spectrographic Analysis. D. V. Evans and D. Johnston. (*Metallurgia*, 1955, **51**, May, 261–262). A brief account is given of a spectrographic technique in which the alloy specimen is dissolved in dilute nitric acid and the solution sprayed via an atomizer across the spark gap of the electrodes. Satisfactory results have been obtained with brasses.—B. G. B.

Metallurgical Analysis: Some Physico-Chemical Considerations of Precipitation Reactions and Electrolytic Separations. C. E. A. Shanahan. (*Metallurgia*, 1955, **51**, May, 255–261). The physico-chemical principles of reactions involving the formation of precipitates (e.g. sulphides, hydroxides) are discussed. Electrolytic separation techniques are considered and the use of polarographic analysis is briefly explained.

Direct Spectral Analysis, Using Displacement. R. Breckpot and M. C. De Clippeleir. (*17th Cong. G.A.M.S.*, 1954, June 23–25, 179–187). An improved spectrograph is described with reference to its use for analysing ferrous alloys. Dispersion and resolution are improved, allowing the separation of two rays of $\Delta\pi = 0{\cdot}033$Å. Working conditions for P, Cu, Mn, C and Si in steel are given.—R. P.

On the Analytical Method of Determining Nitrogen Monoxide in Coke-Oven and Town Gas. V. Spanilý. (*Palivo*, 1955, **35**, (6), 163–167). [In Czech].

Heat-Transfer Used as Parameter in Measuring the Composition of Gas Mixtures. P. Gilli. (*BWK*, 1955, **7**, July, 308–309). Differences in the thermal conductivity are shown to lend themselves for measuring CO_2 contents in flue gases, moisture contents of steam, and the dew-point. An instrument designed by the author, and applied in practice, is discussed.—P. F.

Determination of Oxygen in Certain Gases. Improved Winkler Method. L. Silverman and W. Bradshaw. (*Anal. Chim. Acta*, 1955, **12**, June, 526–541). An improved method is described for the determination of oxygen in some gases, in concentrations down to 1 p.p.m., with a mean standard deviation of ± 0·33 p.p.m.—L. D. H.

Contribution to the Study of Sampling, Reduction and Analysis of Combustible Minerals. R. Rieffel and J. Danquigny. (*Chaleur et Ind.*, 1954, **35**, Oct., 277–302; 1955, **36**, Mar., 73–94; Apr., 126–136). The first section studies the analysis of a sample, by the same and by different operators, and the errors involved. Next, reduction of a sample, by manual or mechanical methods, and the factors giving rise to errors, are examined. The third section deals with the actual sampling of cokes and coals from wagons, trains of wagons, and barges. Sampling standards of France, Germany, U.S.A., and U.K. are compared.—T. E. D.

Automatic Gas Analysis. D. Saint-James. (*Chim. Analy.*, 1955, **37**, Feb., 31–38). Methods described include an automatic Orsat apparatus, an apparatus which estimates the gas by absorption in an electrolyte and measurement of conductivity, analysers determining gases by differences in thermal conductivity, paramagnetic analysers, and infrared absorptiometers.—T. E. D.

Recent Advances in the Analytical Chemistry of Silicate Materials. H. Bennett. (*Trans. Brit. Ceram. Soc.*, 1955, **54**, June, 319–332). Methods available or likely for chemical analysis in the ceramic industry are surveyed. 'Traditional' and 'instrumental' methods are considered in relation to their fields of application.—D. L. C. P.

INDUSTRIAL USES AND APPLICATIONS

The Use of Creep-Resisting Steels in Steam Power Plants. H. W. Kirkby. (*Metallurgia*, 1955, **51**, Apr., 165–170). The selection of steels for steam piping in power stations is discussed and consideration is given to the relative merits of ferritic and austenitic steels. Steels for the construction of turbine rotors, piping, turbine blades and bolts are reviewed in connection with the tendency for higher operating temperatures (33 references).—B. G. B.

Production of the Aircraft Turbine Engine. K. T. Fulton. (*Aeroplane*, 1955, **88**, Mar., 301–329). Under five headings the main materials and processes of manufacture of gas turbines are discussed: the production of blades, compressor and turbine discs, compressor casings, circular and ring parts and sheet metal and other parts. Full details of the alloy steels used and their machining are given.—B. G. B.

Sandwich Structures for High Temperature Service. (*Engineer*, 1955, **199**, Apr. 1, 466–468). This article reviews a number of representative applications of honeycomb structures of all metal construction which are believed to constitute the best available solution to many aircraft structural problems. The material gives a high strength to weight ratio and has good cooling properties.—M. D. J. B.

Heat-Treated Structural Steels. P. O. Björkman. (*Tekn. Tidskr.*, 1955, **85**, Apr. 19, 365–371). [In Swedish]. The author points out that many industries use highly alloyed steels for purposes where low-alloy steels would suffice, and discusses factors governing the choice of steel. He compares

the mechanical properties of mild and alloyed steels, and gives brief details of tests normally employed for determining steel characteristics.—G. G. K.

Non-Ferrous Metals and Stainless Steels in the Coke Oven By-Product Industry. G. A. Dummett and G. H. Botham. (*Yearbook Coke Oven Man. Assoc.*, 1955, 304–329). Copper and copper alloys are used where high conductivity is required. They cannot be used in the presence of ammonia or in oxidizing conditions. The performance of nickel in the presence of hot concentrated alkalies, and of stainless steels of various types in the presence of acids, is outlined. Intergranular corrosion and stress corrosion cracking are described. Corrosion by phenols and tar acids is mentioned, and results of corrosion tests on still effluent and crude liquor are given with a discussion. (21 references).—T. E. D.

Stainless Steels. (*Usine Nouvelle*, 1955, **11**, Mar. 17, 35). The history of stainless steels from about 1905 onwards is briefly surveyed, and the properties of the three principal classes, namely martensitic, ferritic and austenitic steels, are given. Their uses in the chemical and pharmaceutical industries, in dyeing, bleaching and tanning, and in all branches of the foodstuff industry are indicated.—T. E. D.

Cast Steel and Welded Construction. (*Usine Nouvelle*, 1954, **10**, Dec. 23, 28–33). This is a summary of the characteristics and properties of items produced by the following methods: (i) by casting, (ii) from rolled components welded together, (iii) from cast parts welded to rolled, forged or stamped parts, (iv) from components cast and then welded together.—T. E. D.

Evolution of Stainless Steels and Their Uses. J. Dedieu. (*Usine Nouvelle*, Special No., Spring, 1955, 13–19). The development of stainless steel from the original chrome steels to the present range of martensitic, ferritic and austenitic stainless steels is surveyed. The roles of the alloying elements are discussed, and the effects of chromium and nickel on the constitution of the steels at various temperatures are shown diagrammatically. Resistance to chemical corrosion and mechanical properties at low and high temperatures are mentioned.—T. E. D.

The Use of Killed Basic-Bessemer Steel for Oil-Field Pipes. E. Baerlecken and W. Scheurer. (*Stahl u. Eisen*, 1955, **75**, June 2, 718–721). The authors report on Izod tests performed on killed basic-Bessemer and open-hearth steels in the annealed and aged conditions as well as on tensile and welding tests. It is shown that killed and carefully denitrided basic-Bessemer steel is equal, and in some respects superior, to open-hearth steel of similar quality for use in oil-field pipes. T. G.

Steels Used in Moulds for Plastics. (*Aciers Fins. Spéc. Franç.*, 1955, July, 32–34).

A Review of Ferrous Wire Qualities Appropriate to Chain Making. P. L. Lewis. (*Frederick Smith and Co., Wire Manufacturers Ltd.*, June, 22, 1955). The principles of manufacture of electrically welded steel chain are explained. The properties required in the steel wire are described by reference to British Standard Specifications and discussed with particular reference to low-temperature and strain-age embrittlement of chains in use. Causes of faulty chain attributable to the wire are instanced. Alloy steel chains are briefly referred to.

Wire-Drawing Alloy Steels. (*Aciers Fins Spéc. Franç.*, 1955, July, 49–52). A number of examples are given showing the use of special steel wires in industry, together with details of the properties and composition of the steels mentioned.

Fabrication of High-Permeability Magnetic Materials. (*Engineer*, 1955, **199**, Feb. 25, 285–288). This article reviews the special properties aimed at and the current processing methods being employed in the manufacture of high-permeability magnetic materials. Current demand has led a number of American manufacturers of laminations and cores to install the specialized rolling, slitting and roller levelling equipment which is described.—M. D. J. B.

A Survey of the Development of Structural Steels. A. Legat. (*Berg. Hüttenmänn. Monatsh.*, 1955, **100**, Jan., 23–27). The evolution of special steels for constructional purposes is traced in broad outline. Specific properties of special purpose steels are referred to in relation to methods of treatment and alloying elements.—E. C.

Construction and Maintenance of Chemical Apparatus Made of 18/8 Stainless Steels. L. Guiton. (*Usine Nouvelle*, 1955, June, 45–49). Types of stainless steels suitable for use in chemical apparatus are given, with some discussion on types of corrosion encountered. The approach to corrosion and to the construction of stainless steel chemical plant that should be taken by chemical engineers, is outlined.

Some Recent Developments in Stainless Steels. J. I. Morley. (*Iron Steel*, 1955, **28**, Apr., 143–146; May, 183–188). The author draws attention to possible new requirements of the aircraft industry, and considers the possibilities of high-strength stainless steels for these and other new applications. Details are given of the properties and metallurgical background of martensitic 11% Cr steels and precipitation-hardening 17% Cr, 7% Ni steels, and consideration is also given to cold-worked stainless steels.—G. F.

Industrial Buildings. (*Acero Energ.*, Special Issue, 1955, May, 29–104). [In Spanish]. The whole of this issue is devoted to a description of the design and technique of construction of industrial buildings. The illustrations are by photographs and data. (19 references).—P. S.

HISTORICAL

History of Gun Tubes. III. Steel for Cannon. P. R. Kosting. (*Metal Progress*, 1955, **67**, Apr., 109–118, 134–138). The transition between cast iron and steel for the construction of guns in America between 1861 and 1888 is described.—B. G. B.

The Early History of Blister-Steel Production. F. M. Ress. (*Stahl u. Eisen*, 1955, **75**, July 28, 978–980). Records recently discovered in the Hauptstaatsarchiv in Munich show that the cementation process was used in 1601 by Antonius Zeller who claimed to have invented it, and that Johann Nussbaum applied it on a commercial basis in Nuremberg in 1602. Details of the process are given.—T. G.

History of Ironmaking. S. K. Nanavati. (*Tisco*, 1955, **2**, July, 89–97). The author traces the development of blast-furnace ironmaking from primitive times and describes briefly the very early Indian and Chinese furnaces. The substitution of charcoal by coke, the introduction of Neilson's hot-blast technique in the early nineteenth century and other technical advances are dealt with in chronological order. The author then describes the evolution of the modern blast-furnace and its ancillaries and concludes by outlining the history of ironmaking in India. (22 references).—L. E. W.

Samuel Schröderstiernas "Notes on the English Iron Industry" (1749). (*Edgar Allen News*, 1954, **33**, Aug., 185–186). A translation, with notes, is given of a Swedish report on the position of iron and steel production and utilization in England around 1749. Concern was caused at the time by the success of production in America, which affected Swedish bar iron exports and the English industry.—D. L. C. P.

The Evolution of Iron and Steel Making in Nivernais. (*Techn. Civil.*, 1954, **3**, (5), 161–170). Progress of the industry in Nivernais is traced from earliest times until the installation of the first French Bessemer converter at Imphy in 1855.

Benjamin Huntsman. E. N. Simons. (*Metallurgia*, 1955, **51**, Apr., 181–184). This is a short biographical account of the activities of Benjamin Huntsman who first produced cast steel.—B. G. B.

ECONOMICS AND STATISTICS

The Upward Trend of the World Steel Industry. F. Grotius. (*Stahl u. Eisen*, 1955, **75**, May 19, 613–618). The present upward trend of the world iron and steel industry is reviewed and the role of each country analysed. The development now taking place in various overseas countries is discussed. The problems connected with the increasing shortage of scrap is briefly dealt with.—T. G.

Conversion and Processing Cost Analysis in Iron and Steel Works as a Guide to Rationalization. K. E. Poppe. (*Stahl u. Eisen*, 1955, **75**, May 19, 633–640). A cost analysis in iron and steel works covering the whole process from smelting to rolling is given. Examples are quoted with particular emphasis to the comparison of works which operate under different conditions.—T. G.

State of the Use of Analytical Assessment of Personal Output in the Iron and Steel Industry at the End of 1954. H. Euler and E. Blome. (*Stahl u. Eisen*, 1955, **75**, May 19, 641–642). The authors show that the application of the scheme of analytical assessment of personal output is steadily increasing in the German iron and steel industry as a basis for the weekly payroll. 80% of the wage earners in the Western German iron and steel industry will be classified in their earnings by this scheme in the near future.—T. G.

The European Steel Market 1954. H. W. A. Waring. (*Stahl u. Eisen*, 1955, **75**, Apr. 21, 445–452). A review of the

present state of steel production in Western Europe is presented. The steel production for 1954 is estimated to reach 69 million tons, the steel consumption in Western Germany amounts to 180 kg. per head of the population compared with 624 kg. in the U.S.A. The author deals in detail with: (*a*) raw materials, (*b*) production, (*c*) subventions, (*d*) standardization of qualities, (*e*) sheet and tube production, (*f*) competition with other materials (non-ferrous metals and plastics), (*g*) technical development, and (*h*) export.—T. G.

Some Special Problems of the Iron and Steel Industry of Yugoslavia. C. Rekar. (*Berg-Hüttenmän. Monatsh.*, 1954, **99**, Nov., 208–219). This is a wide survey of problems in branches of Yugoslav iron and steel production. Deposits of ores of iron and manganese, and of coal are described. Problems arising from the presence of barytes and arsenic are discussed. Metallurgical problems relating to metal and slag composition are considered, including high-alumina slags and methods of desulphurization. Complete materials balances are given for alternative Bessemer and open-hearth steel production—E. C.

Milestones along the Path of the Austrian Iron and Steel Industry since 1855. R. Walzel. (*Berg-Hüttenmänn. Monatsh.*, 1955, **100**, Jan., 6–19). This article is a simplified survey of Austrian industrial development in parallel with political history. The mixture is one of patriotism and expediency, providing little new factual information.—E. C.

Development of the Iron and Steel Industry, 1953 to 1958. (Iron and Steel Board Pamphlet, 1955, 42 pp.). The plan of development of the British Iron and Steel Industry from 1953 to 1958 presented to Parliament by the Iron and Steel Board is reproduced in its entirety. Full details of the developments carried out by individual firms and of estimated demands for products are given.—B. G. B.

The Australian Iron and Steel Industry's Post-War Expansion. (II). N. R. Wills. (*B.H.P. Rev.*, 1955, **32**, Mar., 1–8).

European Coal and Steel Community. J. Goormaghtigh. (*International Conciliation. Carnegie Endowment for International Peace*, 1955, May, No. 503). An account of the origin, ratification and first two years operation of this European project is given.—B. G. B.

MISCELLANEOUS

Industrial Applications of Atomic Energy. A. Tammaro. (*Steel Processing*, 1954, **40**, Nov., 715–722). The uses of radioisotopes in radiographic inspection, thickness and density gauges, and wear studies are briefly described. In connection with the design of plants to produce power from the atom, a brief review is given of various possible types of reactors.

Adhesive Bonding, H. Thielsch. (*Mat. Methods*, 1954, **40**, Nov., 113–128). A comprehensive account is presented with much data in tabular form, of the many types, properties, forms and uses of modern structural adhesives.—P. M. C.

Sample Tests—Note Without Calculation. M. Dumas. (*Mém. Artillerie Fran.*, 1955, **29**, (2), 305–331). Various aspects of the subject are discussed, in the light of previous work by the author on the applications of probability calculations and statistical methods in industry.—T. E. D.

A Review of Iron and Steel Literature for 1954. V. S. Polansky. (*Blast Furn. Steel Plant*, 1955, **43**, May, 508–514, 545).

Oxidation of Iron–Molybdenum and Nickel–Molybdenum Alloys. S. S. Brenner. (*J. Electrochem. Soc.*, 1955, **102**, Jan., 7–15). The oxidation characteristics of Fe–Mo and Ni–Mo alloys containing up to 12·5 and 19·7 atomic % Mo respectively was studied in stationary and flowing atmospheres of oxygen. In both systems catastrophic oxidation did not occur and in addition to nickel or iron oxides, MoO_2 and molybates were formed at temperatures of 800°–1000° C. The oxidation rate of iron was markedly decreased by the addition of molybdenum. Below 3 atomic % Mo the oxidation rate of Ni was only slightly affected; with 3–12 atomic % Mo the rate was increased and with >12 atomic % the rate was again decreased.—A. D. H.

Catastrophic Oxidation of Some Molybdenum Containing Alloys. S. S. Brenner. (*J. Electrochem. Soc.*, 1955, **102**, Jan., 16–21). The oxidation rate of Mo–Fe, Mo–Fe–Ni, Mo–Cr–Fe, and Mo–Cr–Fe–Ni alloys was studied. It is shown that the presence of chromium or nickel promote catastrophic oxidation of Fe–Mo alloys which oxidize normally (preceding abstract), alloys containing sufficient nickel do not oxidize catastrophically and it is suggested that a Mo–Fe–Ni–Cr alloy containing 40% Ni and 20% Mo may be useful industrially. A theory of catastrophic oxidation is proposed.—A. D. H.

Hollow Iron Cathode Discharge as Source for Wavelength and Intensity Standards. H. M. Crosswhite, G. H. Dieke and C. S. Legagneur. (*J. Opt. Soc. Amer.*, 1955, **45**, Apr., 270–280). A sealed-off hollow cathode discharge tube with iron electrodes and rare gas filling is described which will give an iron spectrum with very sharp lines and intensities of high reproducibility. For most purposes a neon filling at about 3 mm pressure is recommended.—E. C. S.

Interferometric Wavelengths of Iron Lines from a Hollow Cathode Discharge. R. W. Stanley and G. H. Dieke. (*J. Opt. Soc. Amer.*, 1955, **45**, Apr., 280–286). Interferometric wavelength measurements (3570Å to 5709Å) are given for 189 iron lines produced by a hollow cathode discharge. The lines are considerably sharper than those of the iron arc.

A High-Temperature X-Ray Goniometer. R. G. Butters and J. G. Parr. (*Canad. J. Techn.*, 1955, **33**, March, 117–121). A unit is described which, used in conjunction with a Phillips high-angle X-ray diffraction goniometer, allows spectrometer plots to be made to a temperature of 1100° C.—E. C. S.

Studies on the Wetting Effect and the Surface Tension of Solids. The Weakening Effect on the Young's Modulus of Quenched Carbon Steel with its Application for Determining the Surface Tension. M. Sato. (*Proc. Japan. Acad.*, 1954, **30**, Dec., 965–969). Young's modulus in a quenched steel is progressively lowered by wetting by liquids of increasing surface tension (methanol, glycerine and distilled water), and the results can be used to obtain the surface tension of the steel (approx. 350 dynes/cm at 18–20° C).—K. E. J.

Professional Status of the Technicians Employed in Iron and Steel Works. B. Weissenberg. (*Stahl u. Eisen*, 1955, **75**, Mar. 10, 281–287). An analysis of the training and education of the technicians employed in the German iron and steel industry.—T. G.

Glass for Protection Against Heat-Radiation. H. G. Wenzel and H. Spitzer. (*Stahl u. Eisen*, 1955, **75**, Feb. 24, 230–231). The authors describe the use of various types of glass for the protection against radiation in iron and steel works. Either reflecting or absorbing glasses are used. Reflecting glasses are expensive and available only in small sizes. Absorbing glasses contain various quantities of iron oxide and are therefore green. Absorption curves for iron-oxide-containing glasses are presented. The authors deal also with protection against radiation from the sun, e.g. in glass roofing.—T. G.

Cost Accounting in the Iron and Steel Industry in the U.S.A. H. Diehgans. (*Stahl u. Eisen*, 1955, **75**, Feb. 24, 210–220). The same author reviews and discusses the cost accounting systems and methods as used in the American iron and steel industry and compares them with those employed in Germany. T. G.

Bibliography and Index on Dynamic Pressure Measurement. W. G. Brombacher and T. W. Lashot. (*Circ. Nat. Bur. Stand.*, No. 558, 1955, Feb., 124 pp.).

BOOK NOTICES

AREND, HEINRICH, and WERNER NEUHAUS. "*Die Härtbarkeit des Stahles.*" 8vo, pp. 260. Illustrated. Essen, 1955: Verlag W. Girardet. (Price DM. 28.80).

This book follows the lines that one would expect from the title and in general it is a good book, but there is no index and this detracts greatly from its value. It is divided into ten chapters, the first being of an historical and introductory nature. The second chapter deals with the physical properties of iron with solid solutions, with iron carbide and special carbides, and with iron–carbon and iron–manganese thermal equilibrium diagrams. Chapter 3 deals with hardenability from a semi-theoretical point of view, whilst Chapter 4 deals with various practical aspects of hardening, i.e. rate of cooling, effect of cross-section, heat conductivity, the nature of the quenching medium, etc. Chapter 5 is devoted to a consideration of hardness, and in this book it seems a little unnecessary and out of place. Chapter 6 gives a résumé of experimental methods used for

the determination of hardenability. In the later chapters the authors deal not only with the influence of the composition of the steel on the hardenability but also with the effect of the precise method of manufacture.

The book concludes with an excellent bibliography of 743 references.—J. FERDINAND KAYSER.

TRINKS, W. "*Industrial Furnaces*" Volume II, Third Edition. 8vo, pp. x + 358. Illustrated. New York, 1955: John Wiley and Sons Inc.; London: Chapman and Hall Ltd. (Price 80*s*.)

Anyone who has to use or design furnaces will welcome this third edition of the second volume of Professor Trinks' well-known book. It is, like all his works, authoritatively written and largely based upon his wide experience—mostly of U.S. practice.

Conditions in the U.S.A. are emphasized throughout, in regard to both fuels and their economic use. For instance, the author states on p. 20, that the utilization of gas is wasteful in U.S. blast-furnace practice, and that no gas-holders are used. Such a statement certainly does not apply to British or Western European conditions.

Likewise, the utilization of waste heat by recuperation and regeneration is given scant treatment; this vast subject, so vital to furnace efficiency, barely occupies seven pages. The reason is, of course, that fuel gases, such as natural gas, are so relatively cheap in the U.S.A. that the capital expenditure on equipment for waste-heat utilization has rarely been worthwhile.

The first three chapters, dealing with fuels, combustion appliances, fuel mixing and proportioning devices, and temperature control, are excellently written and illustrated. Professor Trinks, however, includes in these chapters the corresponding information on electric heating methods, and by comparison the sections on electric heating are neither as competently nor as comprehensively treated. For instance, he omits several well-known methods of reducing or regulating the heat input of an electric furnace to conform to the thermal demand; star-delta and series-parallel switching, the use of an infinitely variable voltage transformer of the simple carbon-brush type, and the use of induction regulators, are not even mentioned.

On p. 165 he refers to a "sufficiently strong transformer"; whether this means one sufficiently robustly constructed, or one with sufficiently high power output, the reader is left in doubt.

Control of furnace atmosphere is very adequately dealt with in Chapter IV, but the author is apparently unaware of the system of scale-free heating, developed in the U.K. a few years ago, using two-stage combustion, the first stage employing oxygen-enriched air.

The next chapter, covering labour-saving devices, describes very graphically almost every method of mechanization known to furnace designers, and is illustrated on almost every page.

Chapter VI is equally interesting, and gives, among other things, comparisons of fuels and types of furnaces, and includes the important subject of methods of applying heat.

In conclusion, the author has added a very useful chapter on safety measures, which does not appear in the previous editions of this book.

The book lacks any bibliography, and the author says in his preface that "a bibliography tells us what was." Many will disagree with this, as a bibliography can refer the reader to contemporaneous publications dealing much more comprehensively with certain aspects of a subject than is possible within the compass of such a text-book. Nevertheless, many useful references are given in the text, which is well written, copiously illustrated, and clearly printed.—A. G. E. ROBIETTE.

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R. W. EVANS, T.D., M.Met., F.I.M.

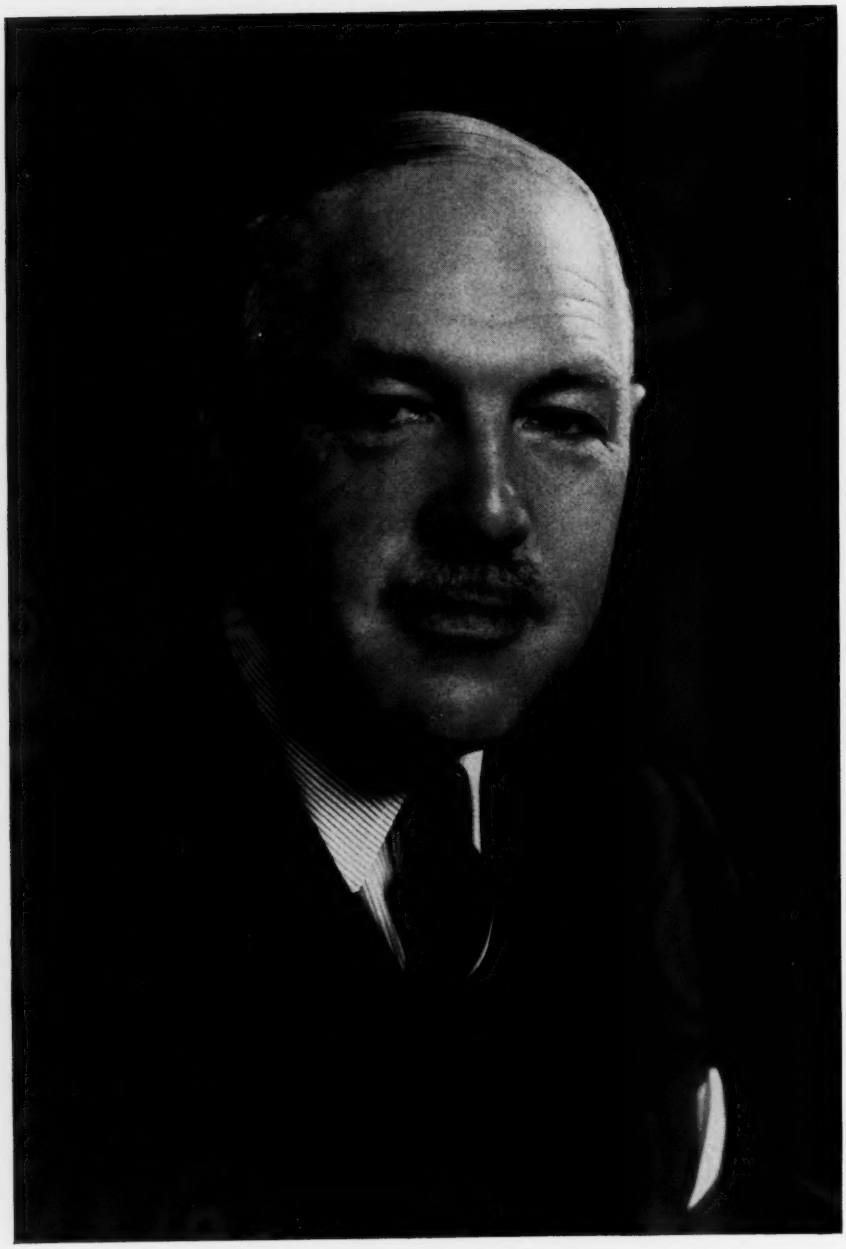
RHYS WILLIAM EVANS was born in 1903 at Merthyr Tydfil, Glam. His family has had connections with the iron and steel industry for over a century; his father, the late Mr. William Evans, General Manager of the Dowlais and Cyfarthfa works and all the collieries of Guest Keen and Nettlefolds Ltd., was President-Elect of The Iron and Steel Institute at the time of his death in 1915.

Mr. Evans was educated at Shrewsbury School and Sheffield University; he graduated B.Met. in 1924 and was awarded the Silver Prize of London City and Guilds in iron and steel manufacture. He entered the Dowlais works of Guest Keen and Nettlefolds Ltd. in the same year as trainee assistant in the Blast-furnaces Department. In 1927 he was transferred to the Bessemer Department as Assistant Manager, and worked there until 1929, when he was appointed Night Manager of the Rail Mill.

On the formation of Guest, Keen and Baldwins Iron and Steel Co., Ltd., in 1934, Mr. Evans was sent to Margam Works, Port Talbot, as Assistant Melting Shop Manager, and became Melting Shop Manager in 1937. He was promoted in 1943 to Superintendent of Melting Shops, which, at that time, included Margam and Port Talbot. Almost immediately Mr. Evans took part in the planning and design of Abbey Melting Shop, which started operations in April, 1951, under his supervision. In December, 1955, he was appointed Assistant General Manager (Operations) of the Steel Division of the Steel Company of Wales Ltd.

During these years Mr. Evans, who joined The Iron and Steel Institute in 1928, contributed a number of papers to the technical press, and he was awarded a Williams Prize for a paper published in the *Journal* in 1945 on "The Heating of Open-hearth Furnaces with Mixed Coke-oven and Blast-furnace Gases." In 1949, Mr. Evans was awarded the degree of Master of Metallurgy, and in 1955 he became a Fellow of The Institution of Metallurgists. He is also a member of The Institute of Physics, the Atomic Scientists Association, the American Institute of Mining and Metallurgical Engineers, and the American Association of Iron and Steel Engineers. In recognition of his service in the Territorial Army from 1929 to the end of the second World War, Mr. Evans was awarded the Territorial Decoration in 1954.

As President of the Swansea and District Metallurgical Society for 1955–1956, Mr. Evans serves as an Honorary Member of Council of The Iron and Steel Institute during his term of office.



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Honorary Member of Council, 1955–1956